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EFFECT OF CONTROLLED ROLLING ON THE CRYSTALLOGRAPHIC TEXTURE AND THE MECHANICAL AND BALLISTIC PROPERTIES OF STEEL ARMOR PLATES

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By Hsun Hu

United States Steel Corporation
Research Laboratory
Monroeville, Pennsylvania 15146

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Foreword

This report was prepared by the Research Laboratory of United States Steel Corporation under U. S. Army Contract No. DAAG46-79-C-0010. The contract was administered by the U. S. Army Materials and Mechanics Research Center, Watertown, Massachusetts, Anthone Zarkades - Contracting Officer's Representative. This is a final report and covers work conducted from January 8, 1979 to January 7, 1980.

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Effect of Controlled Rolling on the Crystallographic Texture and the Mechanical and Ballistic Properties of Steel Armor Plates

by Hsun Hu

Abstract

The scope of the present research included two parts: (1) to further strengthen the (112)+(111) texture in the 5Ni steel (actually a medium-carbon 5Ni-Si-Cu-Mo-V steel) armor by introducing modifications to the previously employed hot-rolling process, and to provide latitude in production of this superior textured armor plate; (2) to investigate the crystallographic texture formation in an austenitic steel system (such as the A-286 alloy) so that textures other than those formed in quenched-and-tempered martensite can be produced and their effects on properties can be studied.

Results indicate that the intensity of the (112)+(111) texture in the 5Ni steel armor can be further increased by controlled rolling with declining temperatures, and that the V_{50} ballistic limits of the 60, 70, and 80 percent rolled plates correlate with the texture-intensity parameter within the scatter band of the earlier data. However, for the 90 percent rolled plates, the V_{50} ballistic limit falls below the scatter band of the correlation. Also, the back-spalling resistance of the heavily rolled plates appears to be lower. These results have been discussed in relation to the changes of other mechanical properties as a consequence of the low rolling and finishing temperatures employed.

Results from the present study on the A-286 alloy indicate that a complete transition of the rolling texture from the copper-type to the brass-type can be effected by decreasing the rolling temperature. A large variety of textures can be produced by appropriate thermal and mechanical processing treatments. These textures can probably be utilized to advantage only in specific applications for sheet materials. The alloy does not appear to be suitable for producing strongly textured thick plates because of the very limited temperature range for hot working to high reductions without concurrent recrystallization.

Introduction

To further extend the study of the effect of crystallographic texture on the mechanical and ballistic properties of high-hardness steel armor plates, the contract with the Army Materials and Mechanics Research Center was renewed for another year in January 1979. The scope of the present research program can be described in two parts.

First, because the beneficial effect of the (112)+(111) texture on the ballistic performance of the quenched-and-tempered 5Ni steel (actually a medium-carbon 5Ni-Si-Cu-Mo-V steel) armor has been well established,[1-3)*] and because of the increased interest in producing armor plates of thinner gages with this texture for improved properties,[4,5)] it appears highly desirable

*See References.

to extend our studies from previous thermomechanical processing treatments involving mainly a final isothermal rolling at 1500°F (816°C) to controlled rolling with declining temperatures. The possibility of finish-rolling at somewhat lower temperatures would further strengthen the texture of the austenite, hence the texture of the martensite. Furthermore, successful application of this controlled rolling with declining temperatures would provide more latitude in production of the textured armor plates, including those of the thin gages.

Second, to investigate the rolling texture formation in a stable-austenitic steel system, it was decided[6] that the A-286 alloy could be used because the alloy is fcc at all temperatures and can be strengthened by precipitation hardening. Consequently, different crystallographic textures other than the (112)+(111) type[1,2] (martensite produced by quenching straightaway-rolled austenite), the ~(110) type[1,7] (martensite produced by quenching cube-textured recrystallized austenite), and the ~(111) type[2,7] (martensite produced by quenching cross-rolled austenite) can be produced; their effects on the mechanical and ballistic properties of the plates can then be studied.

Material and Procedures

Steel Composition and Ingot Dimensions

One 500-pound (227 kg) heat, aimed to the same nominal composition as the steels used in earlier investigations,[1,2] was melted and cast in vacuum in the laboratory. The dimensions of the ingot were also the same as those of the ingots made previously, namely 7 by 12 by 24 inches (180 by 300 by 600 mm). Results of ladle and check analyses are given in Table I. The sample for check analysis was taken from one of the slabs preliminary hot-rolled to the smallest intermediate thickness (1.40 inches or 36 mm) near the top end of the ingot. The present steel matched closely in chemistry with those steels used previously.[1,2,5]

Preliminary Hot Rolling and Preparation of the Intermediate Slabs

After solidification in the mold, the ingot was hot-charged into a preheating furnace at 2250°F (1230°C) and soaked for two hours. The procedures for preliminary hot rolling to slabs of four intermediate thicknesses, namely 5.50, 2.75, 1.85 and 1.40 inches (140, 70, 47, and 36 mm), were the same as those employed in earlier investigations.[1,2] The ingot was first rolled from 7 to 5.50 inches thick. A predetermined length was torch-cut from the bottom end of the slab, and the remaining

portion of the slab was reheated to temperature by returning it to the furnace for about 20 minutes. The piece was then further rolled to the next intermediate thickness (2.75 inches), and another predetermined length was torch-cut from the previous cut end of the rolled slab. The reheating, hot rolling, and torch cutting were repeated when the slab was rolled to 1.85 inches thick, and finally the remaining piece was further rolled to 1.40 inches thick, the smallest thickness of the intermediate slabs.

Following previous practice, all these slabs of various thicknesses were first cooled in air to approximately 800°F (427°C), then cooled slowly in vermiculite to the ambient temperature. These procedures were found to be necessary to eliminate hairline cracks on the cut faces of the slabs. Such cracks frequently induced edge cracking in the finished plates that were isothermally rolled at 1500°F (816°C) to 0.55 inch (14 mm) thick and water-quenched.[2)]

Slabs of the four intermediate thicknesses were cut longitudinally along the centerline of the width into two halves. Thus, each piece became only 6 inches (150 mm) wide for final rolling. A hole 5/32 inch (3.97 mm) in diameter was drilled on the centerline cut face at half thickness and one-third-length position of each piece for accommodating the monitoring thermocouple in final rolling. These preparations were essentially the same as in earlier practices.[1,2)]

Final Controlled Rolling with Declining Temperatures and Subsequent Treatments

For final rolling, the intermediate pieces were reheated at 1700°F (927°C) for two hours. The same rolling schedules employed in earlier investigations[1,2] were used to roll the intermediate pieces, 1.40, 1.85, 2.75, and 5.50 inches thick, to the final 0.55-inch-thick plates in 5, 6, 9, and 18 passes, for a total reduction in thickness of 60, 70, 80, and 90 percent, respectively.* These plates were designated A, B, C, and D for increasing reductions.

In contrast to previous investigations, the temperature of each rolling pass was allowed to decline by a predetermined amount so that rolling started at 1500 to 1600°F (816 to 871°C) and finished at 1300 to 1350°F (704 to 732°C) depending on the total reduction or number of passes. For example, for the 60 percent reduced plates (Plate A), rolling started at 1500°F, finished at 1350°F; for the 70 percent reduced plates (Plate B), rolling started at 1500°F, finished at 1340°F (727°C); for the 80 percent reduced plates (Plate C), rolling started at 1520°F (827°C), finished at 1320°F (716°C); and for the 90 percent reduced plates (Plate D), rolling started at 1600°F, finished at 1300°F. Table II shows the temperatures actually recorded at each pass, together with the reduction schedule employed in the

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*The amount of reduction per pass was mostly around 0.030 inch (0.76 mm), except for the last few passes when lighter reductions were employed to control the final thickness.

final controlled rolling of the plates. The plates were immediately water-spray-quenched to room temperature after the final pass.

The rolled and que[illegible]lates, 0.55 inch thick and 6 inches wide, were subsequently cut into pieces 12 inches (300 mm) long, tempered at 350°F (177°C) for one hour, and cooled in air. These plates were then surface-ground to remove the oxide scale and the decarburized layer before ballistic testing. Specimens for various other mechanical tests, and for microscopic or X-ray examinations, were all prepared from the tempered plates.

Results and Discussion

Texture of the Control-Rolled Plates

The crystallographic textures of the control-rolled plates were examined by X-ray pole figures determined from the midthickness section. Following the same procedures employed previously, both the (110) and (200) pole figures were determined by the Schulz reflection technique up to a tilt angle of 80 degrees from the rolling-plane normal, using filtered MoK_α radiation. Figure 1 shows the (110) pole figures for the plates rolled 60, 70, 80, and 90 percent. The corresponding (200) pole figures of these plates are shown in Figure 2.

As can be seen from these pole figures, the nature and degree of intensity of the textures were similar to those of the plates processed by isothermal rolling at 1500°F in the earlier

investigations.[1,2] However, for the same rolling reductions, the intensities of the textures of the present plates were somewhat higher than those of the previous plates isothermally rolled at 1500°F. This apparently was a consequence of rolling with declining temperatures and with the considerably lower finishing temperatures employed in the present rolling schedules. Similar effects of rolling temperature on texture intensity in quenched 4340 steel armor plate were observed by Zarkades.[8]

Microstructure of the Plates

The microstructures of the control-rolled then quenched-and-tempered plates were examined by optical microscopy and transmission electron microscopy (TEM) at one-million-volt acceleration. Both the longitudinal and the transverse cross sections of the plates were examined. Figure 3 shows the optical micrographs of the plates rolled 60 and 90 percent with declining temperatures then quenched and tempered. As in previous isothermally rolled plates, structural banding was more clearly shown in the longitudinal than in the transverse sections of the specimens. For the 90 percent rolled plates, the banding appeared to be slightly more severe and the bands were more closely spaced than in the earlier plates isothermally rolled 90 percent at 1500°F.[1,2] Other than this, no unusual features could be noticed in the microstructure of the present plates.

The TEM micrographs in Figure 4 show the fine structures of the martensite in the quenched-and-tempered plates rolled 60 and 90 percent. Again, there was very little difference in comparison with the isothermally rolled plates produced in earlier investigations.[1,2] As was observed previously, there was no significant difference in the martensite structure among the plates rolled to 60 and to 90 percent reductions.

In-Plane Tensile Properties

The in-plane tensile properties of the plates were determined by testing specimens 0.25 inch (6.3 mm) in diameter and 1 inch (25 mm) gage length, prepared along three directions in the plane of the plate. These were the longitudinal (L), the diagonal (D), and the transverse (T) directions. Duplicate specimens were tested, and the averaged testing data are summarized in Table III. As can be noted, the strengths and ductility were anisotropic in the plane of the plate. The yield and tensile strengths were higher, whereas the percentage of reduction in area and total elongation were lower, in the transverse direction than in the longitudinal or the diagonal direction. The properties in the diagonal direction were intermediate between those in the longitudinal and the transverse directions but closer to those in the longitudinal direction. The difference between the longitudinal and the transverse properties increased with increasing rolling reduction.

Such in-plane anisotropies were quite similar to those observed earlier in isothermally rolled plates (see Table III in Reference 1). This agreement was expected because the nature of the texture, (112)+(111), is essentially the same for all these rolled and immediately quenched plates. However, the strengths of the present plates were somewhat higher, and the ductility lower, than the corresponding properties of the isothermally rolled plates reported earlier.[1)] It can also be noticed in Table III that the yield strength increased and the total elongation decreased rather consistently with increasing rolling reduction. In those plates isothermally rolled at 1500°F and quenched, such dependence of strength and ductility on rolling reduction was not clearly indicated (see Table III in Reference 1).

These observations suggest that the observed difference in the mechanical properties of the present plates as compared with those of the isothermally rolled (at 1500°F) plates is a consequence of the lowered rolling and finishing temperatures, which would result in higher dislocation contents. Other possible causes, such as the precipitation of second-phase particles (although no direct evidence was indicated by the TEM micrographs), for the observed mechanical properties are at present unclear. Further study will be required to identify all the contributing factors.

Through-Thickness Tensile
Properties of Notched Specimens

As initially suggested by Richmond[9] of the U. S. Steel Research Laboratory, the resistance to spalling at a constant strain rate could be determined for plates by testing the through-thickness tensile strength under constraint conditions, such as with the sharply notched specimen shown in Figure 5, in which the strains $\varepsilon_1 = \varepsilon_2 \approx 0$. In previous investigations a qualitative correlation was observed between the through-thickness tensile strength of such notched specimens and the resistance to back spalling of the ballistically tested plates,[1] or between the through-thickness tensile strength of the notched specimens and the diameter of exit holes of the ballistically tested plates.[7]

Similar tests were conducted on notched specimens of the present plates, and the results are shown in Table IV. As can be noted, the through-thickness tensile strength and total elongation of the notched specimens decreased with increasing rolling reduction. The through-thickness tensile strengths of the present plates were, in fact, appreciably lower than those of earlier plates isothermally rolled at 1500°F, similarly quenched and tempered (see Column 1 of Table XIII in Reference 1), particularly for those plates rolled to high reductions. These results suggested that the resistance to back spalling of the present heavily rolled plates would probably be even lower than that of the earlier plates rolled isothermally at 1500°F to correspondingly

high reductions. This seemed to have been confirmed by the ballistic-testing results for the present plates, as will be reported in a later section.

Charpy Impact Properties

The Charpy V-notch impact properties of the plates were determined by testing at room temperature on duplicate full-size specimens prepared along the longitudinal, diagonal, and transverse directions. The results are summarized in Table V. The impact energy was the highest in the longitudinal direction, lowest in the transverse direction, and intermediate in the diagonal direction. This is in good agreement with the results obtained by Zarkades[10)] on the isothermally rolled (90 percent reduction at 1500°F) plates produced earlier in this Laboratory.[2)] However, the impact energies of the present plates were somewhat lower than the corresponding impact energies of the earlier isothermally rolled plates. For example, the impact energies reported by Zarkades[10)] were 17.5, 15.2, and 11.5 ft-lb (23.7, 20.6, and 15.6 J) for the longitudinal, diagonal, and transverse directions, whereas for the corresponding directions in the present plates (Plate D-1 in Table V), the impact energies were 14.0, 12.5, and 9.8 ft-lb (19.0, 17.0, and 13.3 J), respectively. In comparison with the isothermally rolled plates, the lower impact energies of the present plates were consistent with their higher strengths and lower ductilities. Data in Table V also indicated a consistent dependence of impact energy on rolling reduction.

Ballistic Performance of the Plates

The V_{50} ballistic limits of the present .ates were determined by using 0.50 caliber AP M2 projectiles at zero-degree obliquity. These ballistic limits, together with other pertinent information such as texture type, texture intensity, and the hardness and thickness of test plate, are summarized in Table VI. The V_{50} ballistic limit increased with increasing rolling reduction and the intensity of the (112)+(111) type texture. The texture intensities of the present plates were somewhat higher than those of the correspondingly reduced plates rolled isothermally at 1500°F (compare with Table VIII in Reference 1, and Table III in Reference 2).

To compare the ballistic limits of the present plates with those of the earlier plates as a function of texture intensity, the ballistic limits of the present plates were corrected for thickness to 0.470 inch (a common thickness arbitrarily selected previously for comparison), and added to the plot of Figure 5 in Reference 2. This new plot is shown in the present report as Figure 6. The data points for the present plates are represented by the triangles. As shown, the data points for the A, B, and C plates, which were rolled 60, 70, and 80 percent, respectively, were well within the scatter band. For the plates rolled 90 percent (the D plates), the ballistic limit fell below the scatter band. A tendency for the ballistic limits to level off was also

indicated by the C plates, which were rolled 80 percent with a finishing temperature of 1320°F (716°C).

The back-spalling tendency of these plates can be qualitatively judged by the size and appearance of the exit holes on the back side of the tested plates. Figures 7A to 7D show the photographs of the ballistic-tested plates A-2, B-2, C-2, and D-2 (see Table VI), respectively. The back-spalling tendency of the present plates, particularly those rolled to high reductions, appeared to be somewhat greater than that of earlier plates processed by isothermal rolling to the same reductions. This is in confirmation of the prediction based on the through-thickness tensile strengths of notched specimens of the plates.

These results indicate that the apparent increase in texture intensity obtained by decreasing the finishing temperature to 1300°F (704°C) did not proportionately improve the ballistic limits of the plates. On the basis of the various mechanical properties of the plates observed, such as higher yield and tensile strengths, lower ductility, lower through-thickness tensile strength, and lower impact energy than those of the earlier plates isothermally rolled at 1500°F, the apparent increase in texture intensity may not simply represent an increase in the degree of preferred orientation, but also an actual increase in the dislocation contents. It is also possible that some precipitation may have occurred by rolling and finishing at these low

temperatures, even though no visible precipitation was evident in the TEM micrographs of the specimens (Figure 4). Moreover, in a highly dislocated matrix, such as quenched martensites, it is rather difficult to visually detect the presence of small amounts of fine precipitates.

Summary and Conclusions

In an attempt to further strengthen the (112)+(111) texture and to provide more latitude in production for the 5Ni steel armor, the effect of controlled rolling with declining temperatures (in the range 1600 to 1300°F) on the crystallographic texture and the mechanical and ballistic properties of the plates has been studied. Results indicate that the apparent texture intensities of the present plates were somewhat higher than those of the earlier plates processed by isothermal rolling at 1500°F to corresponding reductions because the temperatures for the majority of rolling passes and the finishing pass were considerably lower than in earlier investigations. The various mechanical properties and the ballistic performance of the present plates were also affected by the low rolling and finishing temperatures. For example, the yield and tensile strengths were higher, whereas the ductility, the through-thickness tensile strength, and the impact energy were lower for the present plates than for the earlier plates.

As was observed in previous investigations, the V_{50} ballistic limit increased with the rolling reduction and the intensity of the (112)+(111) type texture. In comparison with the ballistic performance of earlier plates as a function of the texture-intensity parameter, the V_{50} ballistic limits of the present plates rolled 60, 70, and 80 percent were well within the scatter band of the correlation. However, for the present plates rolled 90 percent, the ballistic limit fell below the scatter band of the correlation. The back-spalling resistance of the present plates, particularly those rolled to high reductions, also appeared to be somewhat inferior to that of the earlier plates rolled isothermally at 1500°F to the same reductions. These results indicate that an apparent increase in texture intensity due to rolling and finishing at considerably lower temperatures may not result in a corresponding improvement in the ballistic properties. As indicated by the differences observed between the various mechanical properties of the present plates and those of the earlier plates rolled isothermally, an actual increase of the dislocation contents and possibly some precipitation hardening may have been effected by the controlled rolling with declining temperatures employed.

Recommendations for Future Work

Results from previous and present investigations on 5Ni steel armor indicated that the (112)+(111) textured plates had ballistic properties substantially superior to those of the random-textured plates. The V_{50} ballistic limit increased with the intensity of the (112)+(111) texture practically linearly provided that the temperature of rolling was within the range 1600 to 1400°F (871 to 760°C). To facilitate the production of such superior textured armor by industrially established processing techniques with few changes required, it would be highly desirable to study the effects of microalloying on the retardation of recrystallization in the 5Ni steel during hot rolling so that the steel can be processed at higher temperatures to high reductions without concurrent recrystallization. A very strong texture would thus be developed in the deformed austenite, hence in the quenched-and-tempered martensite. Hot rolling and finishing at higher temperatures would facilitate commercial production of such textured armor plates even for the thin gages.

Research Part II (A-286 Alloy)

Alloy Composition and Ingot Dimensions

One 500-pound heat, aimed to the nominal composition of the A-286 alloy given in the literature,[11,12] was vacuum-melted and cast in the melt shop of the Laboratory. The ingot, 7 by 12 by 24 inches (180 by 300 by 600 mm), the same as the 5Ni steel ingot, was hot-charged into a preheating furnace at 2000°F (1093°C), and soaked at temper.ture for 2 hours. The ingot was first hot-rolled from 7 to 5.50 inches (180 to 140 mm) thick. After it was cooled to room temperature, a piece abc.it 8 inches (203 mm) in length was cut from the bottom end of the ingot. The remainder of the rolled ingot was reheated to temperature and hot-rolled to 2.75 inches (70 mm) thick. Another piece about 14 inches (356 mm) in length was cut from the previously cut end of the slab. The process of reheating, hot rolling, and cutting was repeated until four slabs having the intermediate thicknesses 5.50, 2.75, 1.85, and 1.40 inches (140, 70, 47, and 36 mm) were obtained.

After the hot top was cut off and discarded and a small slice was taken for check analysis from the 1.40-inch-thick slab, the remaining 1.40-inch-thick material was used for preliminary studies of the microstructural and textural characteristics of the alloy after various thermomechanical processing treatments. Appropriate processing of the intermediate slabs to the final 1/2-inch-thick armor plates with desired crystallographic textures

and high strength depends on the results of these preliminary studies.

The chemical composition of the alloy, as shown by the ladle and check analyses, is given in Table VII. Except for the nitrogen concentration, which is higher than originally aimed (0.005%), the concentrations of all other alloying elements are within the aim range.

Solution Treatment and Microstructure

The microstructure of the hot-rolled slab that was cooled in air consisted of recrystallized grains with annealing twins. There were some coarse (2 to 5 μm in diameter), gray-etching, second-phase particles, believed to be titanium carbide (TiC), because X-ray energy spectra of these particles examined in the SEM (scanning electron microscopy) indicated high titanium contents, and because titanium nitride (TiN) can easily be recognized microscopically by its characteristic angular shape and reddish orange color. These TiC particles could not be eliminated by solution treatments at temperatures as high as 2300°F (1260°C). Small specimens annealed for one hour at various temperatures, ranging from 1800 to 2300°F (982 to 1260°C), and quenched in water showed little difference in hardness, as can be seen in Table VIII. The microstructures of the specimens solution-treated at 2300°F and at 1800°F showed big differences in grain size, but only minor differences in the amount and size of the second-phase particles, Figure 8.

These results suggested that a solution treatment by annealing at 1800°F for 1 hour and quenching should be adequate for the dissolution of the gamma prime (γ') phase $Ni_3(Ti,Al)$ or the eta (η) phase Ni_3Ti—the principal second phase for precipitation strengthening in the A-286 alloy. Furthermore, a smaller penultimate grain size (by solution annealing at a lower temperature) would be more desirable for the development of a strong texture by subsequent thermomechanical processing than a coarser grained material.

Rolling Textures and Their Dependence on Temperature

To provide necessary information for appropriate thermomechanical processing of the 1/2-inch-thick texture plates, the textural behavior of the alloy was studied extensively by using small specimens. Rectangular blocks, approximately 3/4 inch (19 mm) thick by 1-1/4 inches (32 mm) wide by 1-1/2 inches (38 mm) long, were first solution-treated by annealing at 1800°F (982°C) for 1 hour and quenched. The blocks were then isothermally rolled (by reheating to temperature after each pass) to 90 percent reduction in thickness at various temperatures ranging from 1800 to -100°F (982 to -73°C). In a manner similar to the textural behavior of various fcc metals[13)] and alloys,[13)] including the Type 304 stainless steels,[13,14)] the rolling texture of the A-286 alloy changed gradually from the copper-type to the brass-type with decreasing temperature. A completely brass-type texture was

produced at -100°F, by using dry-iced acetone as the cooling bath. In comparison with the Type 304 stainless steels,[14] the temperature for a complete texture transition from the copper to the brass type is much lower in the A-286 alloy (-100°F vs 392°F or 200°C), suggesting that the stacking-fault energy of the A-286 alloy is considerably higher than that of the Type 304 stainless steels.

The temperature dependence of the rolling-texture transition in the A-286 alloy can be described as follows:

1. Rolling at 1800°F (982°C) or even at 1700°F (927°C) causes concurrent recrystallization, resulting in nearly random or very weak textures. Figure 9A shows the microstructure of the specimen rolled 90 percent at 1700°F. The textures of the specimens rolled at 1800°F and at 1700°F are shown by the (111) pole figures in Figure 10, A and B, respectively. As can be noted, the texture of the specimen rolled at 1700°F is slightly stronger than that of the specimen rolled at 1800°F. The observed difference in the degree of texture intensity between these two specimens, both being completely recrystallized, can probably be understood on the basis of the amount of rolling deformation, or number of passes necessary for recrystallization to occur at the temperature of processing. It can also be observed that considerable grain refinement has occurred during isothermal rolling of the specimen at 1700°F (compare the grain size of the solution-treated specimen, Figure 8B, and that of the specimen rolled at 1700°F, Figure 9A).

2. Rolling at temperatures in the range 1500 to above 900°F (816 to above 482°C) produced a strong texture of the copper type. Rolling in the temperature range 1400 to 1300°F (760 to 704°C), even for the small-size specimens, should be avoided because rapidly increased hardening due to precipitation causes extensive cracking of the strip. As shown by the (111) pole figures in Figure 11, the texture of the specimen rolled at 1500°F (Figure 11A) can be described very well by the ideal orientations of the copper type, namely, $\sim\{123\}\langle 412\rangle$, $\{110\}\langle 112\rangle$, and $\{112\}\langle 111\rangle$. When the temperature of rolling is decreased to 900°F, although the same ideal orientations can approximately describe the texture, a substantial broadening of the orientation spread is clearly indicated (Figure 11B). The microstructures of these specimens are shown in Figure 9B (rolled at 1500°F) and 9C (rolled at 900°F).

3. For rolling temperatures in the range 900 to 500°F (482 to 260°C), a gradual transition of the texture from the copper type to the brass type becomes increasingly evident. The textures of the specimens rolled at 700°F (371°C) and at 500°F (260°C) are shown in Figure 12, A and B, respectively.

4. Such texture transition is nearly complete in the specimen rolled at approximately 80°F (27°C), as shown by the (111) and (200) pole figures in Figure 13. Although the central portion of the (111) pole figure has not shown the splitted intensity maxima to represent the $\{110\}\langle 112\rangle$ texture, this ideal orientation is clearly indicated in the (200) pole figure. When the rolling

temperature is further decreased to -100°F (-73°C) by using dry-iced acetone as the cooling bath, a completely brass-type texture is produced, as shown in Figure 14. The microstructure of this specimen, Figure 9D, shows the presence of numerous very finely spaced deformation bands.

Recrystallization Textures of Isothermally Rolled Strips

Annealing the various isothermally rolled specimens at 2000°F (1093°C) for 1 hour caused complete recrystallization; and for those already recrystallized during rolling, there was considerable grain growth. Selected specimens were also given a recrystallization anneal at 1800°F (982°C) for comparison with the specimens annealed at 2000°F. The recrystallization textures were essentially the same upon annealing at both of these temperatures, except for the fact that the texture was a little sharper in the specimen annealed at the higher temperature than at the lower temperature. The recrystallization or annealing textures of the various isothermally rolled strips can be summarized as follows:

1. For the already recrystallized strips (rolled at 1800 and 1700°F) having nearly random or very weak textures, annealing at 2000°F either further randomized or slightly strengthened the texture as a consequence of additional grain growth. This is shown by the (111) pole figures in Figure 15.

2. For the specimen rolled 90 percent at 1500°F having a strong texture of the copper type (Figure 11A), the recrystallization texture is {111}<110> + {001}<110>, as shown by the (111) and (200) pole figures in Figure 16. This is in contrast to the strong cube texture normally observed in other fcc metals or alloys.[13] Such textural behavior upon recrystallization is presumably a consequence of the dissolution or precipitation of the dispersed second-phase particles in the specimen during the recrystallization anneal. Similar effects have been observed in other fcc alloys.[15]

3. For specimens rolled at 900 to 500°F (482 to 260°C) and having the transitional rolling textures (Figures 11B, 12A, and 12B), the recrystallization texture is the {112}<111> type, as shown in Figure 17 by the (111) and (200) pole figures of the specimen recrystallized by annealing at 1800°F (982°C) for 1 hour. Annealing at 2000°F (1093°C) for 1 hour produced the same, but somewhat sharper, recrystallization texture. As shown by the pole figures in Figure 18, the same {112}<111> type recrystallization textures were produced by annealing the specimens rolled at 700°F (371°C) and 500°F (260°C) at 2000°F (1093°C) for one hour.

4. For the specimens rolled 90 percent at approximately 80°F (27°C) and at -100°F (-73°C), having a brass-type rolling texture (Figures 13 and 14), the recrystallization texture is {110}<112> + {110}<001>, as shown in Figure 19. In contrast to the more complex

recrystallization textures of brass, or of other fcc metals or alloys that develop a brass-type rolling texture,[13] the brass-type rolling texture in A-286 alloy was retained upon recrystallization or recrystallization in situ. For the specimen rolled 90 percent at ∿80°F and annealed at 2000°F (1093°C), the recrystallization texture was practically the same as those shown in Figure 19.

Precipitation Hardening in Rolled and Recrystallized Specimens

The aging treatment commonly employed for precipitation strengthening in the A-286 alloy is to anneal the specimen at 1300 to 1400°F (704 to 760°C) for 12 to 16 hours and to cool in air.[12] Depending on the Ti/Al ratio, the precipitated phase is either the gamma prime (γ') phase $Ni_3(Ti,Al)$ or the eta (η) phase Ni_3Ti, as a high Ti/Al ratio indicates a strong tendency for the gamma prime phase to transform into the eta phase after long periods of thermal aging.[16] For the specimens rolled 90 percent at 1500°F (816°C), overaging occurred upon annealing at 1400°F (760°C) for about four hours, whereas aging at 1300°F (704°C) increased the hardness, appreciably in the first hour, but with only minor variations during further prolonged aging periods. These results are shown in Table IX.

Specimens rolled 90 percent at 1500°F (816°C) and recrystallized at 2000°F (1093°C) had a low hardness corresponding to the solution-treated condition. Table X shows the substantial

increases in hardness upon aging at 1300°F (704°C). These data indicate that the hardness reached a maximum after aging 16 to 32 hours. The microstructures of a recrystallized, and of a recrystallized and aged, specimen are shown in Figure 20. There is no great difference visible between the optical micrographs of these two specimens, except for the faintness of the grain and the twin boundaries. The response to etching is drastically different between the specimens in the solution-treated (by the recrystallization anneal) and in the aged conditions. The texture is essentially unchanged after the aging treatment.

Final Rolling of the Slabs

On the basis of the information obtained from the extensive studies of the textural behavior in relation to the thermomechanical processing treatments of the alloy, isothermal rolling at 1500°F (816°C) appeared to be the only sensible possibility for final rolling the intermediate slabs to the 1/2-inch-thick plates. For slabs of such intermediate thicknesses, 1.85, 2.75, and 5.50 inches (47, 70, and 140 mm), even at this temperature (1500°F) isothermal rolling to high reductions poses great difficulties. On a trial basis, we conducted the final rolling as follows.

The slabs were reheated to 1800°F (982°C) and soaked at temperature for two hours. The two 1.85-inch-thick slabs to be rolled 70 percent to 0.55-inch-thick plates (B-1 and B-2 plates) were processed first. For the B-1 plate, the slab was rolled with a start-rolling temperature of 1550°F (843°C), and finish-rolled

at 1380°F (749°C) in 8 passes. The plate was water-spray-quenched after the final pass. To raise the finishing temperature to above 1400°F (760°C) so that the effect of precipitation hardening could largely be avoided, the B-2 plate was rolled to the final thickness also in 8 passes, starting at 1600°F (871°C), finishing at 1420°F (771°C), and similarly quenched. It was noticed that some edge cracking occurred in both plates. This seemingly partial success in the final rolling of the B-plates (70% reduction) led to the trial rolling of the next thicker slabs (2.75 inches or 70 mm thick) to be rolled 80 percent to the final 0.55-inch-thick plates (C-1 and C-2 plates). The start-rolling temperature for the C-1 plate was 1600°F (871°C). On the second pass, the tail portion of the piece was split open widely from the midthickness plane of the plate. The same failure occurred on the C-2 plate in the third pass. The final rolling operation was then stopped.

Upon close examination of the finished B-1 and B-2 plates, it was discovered that besides the numerous cracks on the edges, both plates were cracked extensively from the thermocouple hole (drilled from one of the side faces of the slab to the geometric center of the slab) along the midthickness plane of the rolled plate. Consequently, the plates were all scrapped. The A-286 alloy obviously has to be hot-worked at much higher temperatures for plates of these thicknesses. However, a strong texture will not be produced during such hot working because of concurrent recrystallization.

A small piece was cut out from the scrapped B-1 plate material, and the texture was examined at approximately the mid-thickness section of the specimen. As shown by both the (111) and (200) pole figures in Figure 21, the 70 percent rolled plate had developed a strong copper-type texture, which is in excellent agreement with the rolling texture of the small-size specimen rolled 50 percent isothermally at 1500°F (Figure 11A). The intensity of the texture of the B-plate is, in fact, somewhat stronger than that of the small-size specimen. This is obviously due to the fact that the B plate was actually rolled with a considerably lower finishing temperature and without intermediate anneals between passes.

Summary and Recommendations

The textural behavior in relation to the thermomechanical processing treatments of the A-286 alloy has been studied extensively with small-size specimens. The textures of the as-rolled (at 90% reduction) and of the subsequently recrystallized specimens were examined along with the hardness and microstructures of the processed specimens. Results show that isothermal rolling at 1700°F (927°C) and higher temperatures caused concurrent recrystallication and resulted in very weak textures. Rolling at 1500°F (816°C) and lower temperatures down to above 900°F (482°C) produced a strong texture of the copper type. For rolling temperatures in the range 900 to 500°F (482 to 260°C) indications of a gradual transition of the texture from the copper-type to the brass-type

became increasingly evident. Such texture transition was nearly complete in the specimen rolled at $\sim$80°F (27°C), and a brass-type texture was produced by rolling at -100°F (-73°C).

Annealing the various isothermally rolled specimens at 2000 or 1800°F (1093 or 982°C) for 1 hour produced recrystallization texture of three main types: from the copper-type rolling texture (rolled at 1500 to above 900°F), the recrystallization texture was $\{111\}\langle 110\rangle + \{001\}\langle 110\rangle$; from the transitional rolling textures (rolled at 900 to 500°F), the $\{112\}\langle 111\rangle$; and from the brass-type rolling texture (rolled at $\sim$80 or -100°F), the $\{110\}\langle 112\rangle + \{110\}\langle 001\rangle$. These recrystallization textures which are distinctly different from the recrystallization textures normally observed in common fcc metals or alloys, are developed presumably under the influence of dissolution or precipitation of finely dispersed second-phase particles during the recrystallization anneals. The alloy can be substantially strengthened by precipitation hardening upon aging at 1300 to 1400°F (704 to 760°C); the texture of the specimen is essentially unchanged by such aging treatments.

A wide variety of textures can be produced in the alloy by appropriate thermal and mechanical processing treatments, and these textures can probably be utilized to advantage in specific applications. However, these possibilities can only be suitably applied to sheet materials, for which heavy rolling reductions can be employed at relatively low temperatures without encountering

the effects of precipitation hardening. The alloy does not appear to be suitable for producing strong textures in thick plates because of the very limited temperature range for hot working to high reductions.

Acknowledgement

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References

1. H. Hu, G. R. Speich, and R. L. Miller: "Effect of Crystallographic Texture, Retained Austenite, and Austenite Grain Size on the Mechanical and Ballistic Properties of Steel Armor Plates," AMMRC CTR 76-22, July 1976, Contract No. DAAG46-75-C-0094.

2. H. Hu: "Studies of Texture Development in Steel Armor Plates," AMMRC CTR 77-19, July 1977, Contract No. DAAG46-77-C-0014.

3. A. K. Ghosh and N. E. Paton: "Deformation of Textured Steels," AMMRC TR 78-40, September 1978, Contract No. DAAG46-77-C-0054.

4. Contract Research Meeting on Armor Plate Developments, Army Materials and Mechanics Research Center, Watertown, MA, May 19, 1978.

5. H. Hu: "Laboratory Processing of 1/4-Inch-Thick Steel Armor Plates with a Strong (112)+(111) Texture," Report to AMMRC, October 1978, Contract No. DAAG46-78-M-1074.

6. Contract Research Meeting on Armor Plate Developments, Army Materials and Mechanics Research Center, held at Rockwell International Science Center, Thousand Oaks, CA, January 12, 1979.

7. H. Hu: "Effect of Crystallographic Texture on the Mechanical and Ballistic Properties of Steel Armor," AMMRC TR 78-39, July 1978, Contract No. DAAG46-77-C-0037.

8. A. Zarkades: Private Communication, Army Materials and Mechanics Research Center.

9. O. Richmond: "Some Suggestions for Tensile Specimens to Measure Splitting and Spalling Resistance," U. S. Steel Research Memorandum, October 22, 1975.

10. A Zarkades, "Improved High Strength Armor Steel Through Texturing," Army Materials and Mechanics Research Center, AMMRC TR 79-52, September 1979.

11. <u>Metals Handbook</u>, Vol. 1, Properties and Selection of Metals, ASM, 1961, p. 467.

12. <u>Alloy Digest</u>, June 1977, SS-340, Pyromet A-286 (Heat and Corrosion-Resistant Alloy).

13. H. Hu: "Texture of Metals," Texture, 1974, Vol. 1, No. 4, pp. 233-258.

14. S. R. Goodman and H. Hu: "Texture Transition in Austenitic Steels," Trans. TMS-AIME, 1964, Vol. 230, pp. 1413-1419.

15. E. Hornbogen and H. Kreye: "Influence of Second Phase Particles on Recrystallization," Textures in Research and Practice, Eds. J. Grewen and G. Wassermann (Springer-Verlag, Berlin, 1969), pp. 274-311.

16. J. R. Mihalisin and R. F. Decker: "Phase Transformations in Nickel-Rich Nickel-Titanium-Aluminum Alloy," Trans. TMS-AIME, Vol. 218, 1960, pp. 507-515.

Table I

Chemical Composition of the 5Ni Armor Steel in Weight Percent

Heat No.	C	Mn	P	S	Si	Cu	Ni	Mo	V	Al	N	O ppm	
7961-3019 (or 919)	0.40	0.60	0.004	0.004	1.33	1.00	5.72	0.48	0.09	0.06	0.005	54	(Ladle Analysis)
"	0.40	0.59	0.004	0.004	1.33	1.02	5.73	0.48	0.09	0.05	0.005	29	(Check Analysis)*

*Sample taken from the 1.40-inch-thick slab near the top end of the ingot.

Table II

Temperature and Reduction Schedule for Final Hot Rolling of Steel Armor Plates

	Plate Designation 7961-8019 (or 919)							
	D		C		B		A	
	Intermediate Slab Thickness, in.							
	5.50		2.75		1.85		1.40	
	Total Reduction, %							
	90		80		70		60	
Pass No.	Temperature, °F	Thickness, in.	Temperature, °F	Thickness, in.	Temperature, °F	Thickness, in.	Temperature, °F	Thickness, in.
1	1600	5.25	1520	2.45	1500	1.55	1500	1.12
2	1580	4.95	1500	2.15	1460	1.25	1460	0.88
3	1560	4.65	1480	1.85	1420	0.95	1420	0.71
4	1540	4.35	1460	1.55	1380	0.70	1380	0.61
5	1520	4.05	1440	1.25	1360	0.61	1350	0.55
6	1500	3.75	1400	0.95	1340	0.55		
7	1490	3.45	1360	0.70				
8	1480	3.15	1340	0.61				
9	1470	2.75	1320	0.55				
10	1460	2.45						
11	1450	2.15						
12	1440	1.85						
13	1420	1.55						
14	1400	1.25						
15	1380	0.95						
16	1360	0.70						
17	1340	0.61						
18	1300	0.55						

Conversion Factors

Deg. C = (Deg. F - 32) x 5/9

1 inch = 25.4 mm

Table III

Tensile Properties of Steel Armor Plates Control-Rolled with Declining Temperatures to Various Reductions, Immediately Quenched, and Subsequently Tempered

Plate Designation	Hot Rolling Reduction, %	Yield Strength (0.2% Offset), ksi (MPa)			Tensile Strength, ksi (MPa)			Reduction in Area, %			Total Elongation, %		
		L	D	T	L	D	T	L	D	T	L	D	T
7961-8019													
A-1	60	228.2 (1573)	225.6 (1556)	229.3 (1581)	314.0 (2165)	316.5 (2182)	324.1 (2235)	45.3	45.8	30.9	12.9	12.0	10.0
B-1	70	231.4 (1596)	2[illegible]8.0 (1641)	245.5 (1693)	313.5 (2162)	315.4 (2175)	323.6 (2231)	51.2	43.1	30.9	12.9	12.3	10.0
C-1	80	236.[illegible] (1628)	239.8 (1653)	253.6 (1749)	310.5 (2141)	316.0 (2179)	327.7 (2259)	45.8	41.7	30.2	12.0	11.9	9.0
D-1	90	247.2 (1704)	250.0 (1724)	268.0 (1848)	313.1 (2159)	316.5 (2182)	333.4 (2299)	44.4	42.9	29.3	12.0	11.9	8.0

Results represent the averaged values of duplicate specimens tested for each single plate.

Table IV

Through-Thickness Tensile Properties of Notched Specimens of Steel Armor Plates Control-Rolled with Declining Temperatures to Various Reductions, Immediately Quenched, and Subsequently Tempered

Plate Designation	Hot Rolling Reduction, %	Tensile Strength of Notched Specimens ksi	(MPa)	Total Elongation, %
7961-8019 (or 919)				
A-1	60	421.1	(2903)	0.71
B-1	70	397.3	(2739)	0.40
C-1	80	323.1	(2228)	0.36
D-1	90	292.8	(2019)	0.25

Results represent the averaged values of duplicate specimens tested for each single plate.

Table V

Charpy V-Notch Impact Properties of Steel Armor Plates Control-Rolled with Declining Temperatures to Various Reductions, Immediately Quenched, and Subsequently Tempered

Plate Designation	Hot Rolling Reduction, %	Impact Energy L ft-lb	L (J)	D ft-lb	D (J)	T ft-lb	T (J)
7961-8019 (or 919)							
A-1	60	19.0	(25.8)	15.6	(21.2)	11.5	(15.6)
B-1	70	17.2	(23.3)	15.4	(20.9)	12.5	(17.0)
C-1	80	17.0	(23.1)	14.5	(19.7)	12.0	(16.3)
D-1	90	14.0	(19.0)	12.5	(17.0)	9.8	(13.3)

Results represent the averaged values of duplicate specimens tested for each single plate.

Table VI

Ballistic Performance of Steel Armor Plates Control-Rolled with Declining Temperatures to Various Reductions, Immediately Quenched, and Subsequently Tempered

Plate Designation	Hot Rolling Reduction, %	Texture Type	Texture Intensity	Test Plate Hardness, RC	Test Plate Thickness, in.	Ballistic Limit, V_{50} fps
7961-8019 (or 919)						
A-1	60		4.40	54.0	0.504	Increasing
A-2			-	52.3	0.502	
				<53.2>	<0.503>	
B-1	70	(112)	5.30	54.0	0.500	
B-2			-	52.6	0.501	
		+		<53.3>	<0.500>	
C-1	80	(111)	6.90	55.7	0.498	
C-2			-	55.3	0.500	
				<55.5>	<0.499>	
D-1	90		9.50	55.8	0.498	
D-2			-	55.5	0.500	
				<55.7>	<0.499>	↓

Numbers in < > represent averaged values.

Conversion Factors

1 inch = 25.4 mm
1 fps = 0.328 mps

Table VII

Chemical Composition of A-286 Alloy in Weight Percent

Heat No.	C	Mn	P	S	Si	Ni	Cr	Mo	V	Ti	Al	N	O(ppm)	
7961-8041 (or 941)	0.08	1.38	0.008	0.006	0.56	25.30	14.91	1.29	0.34	2.04	0.17	0.008	102	(1)
"	0.08	1.38	0.005	0.005	0.51	25.44	14.95	1.28	0.33	2.02	0.23	0.008	34	(2)

(1) Ladle analysis

(2) Check analysis of the slab

Table VIII

Hardness of Specimens Solution-Treated by Annealing for 1 Hour at Various Temperatures and Quenching in Water

Annealing Temperature, °F	Hardness DPH
1800 (982°C)	135
1900 (1038°C)	132
2000 (1093°C)	132
2100 (1149°C)	128
2200 (1204°C)	129
2300 (1260°C)	130

Table IX

Precipitation Hardening of the Specimens Rolled 90 Percent at 1500°F (816°C), Then Aged for Various Times at Temperatures Indicated

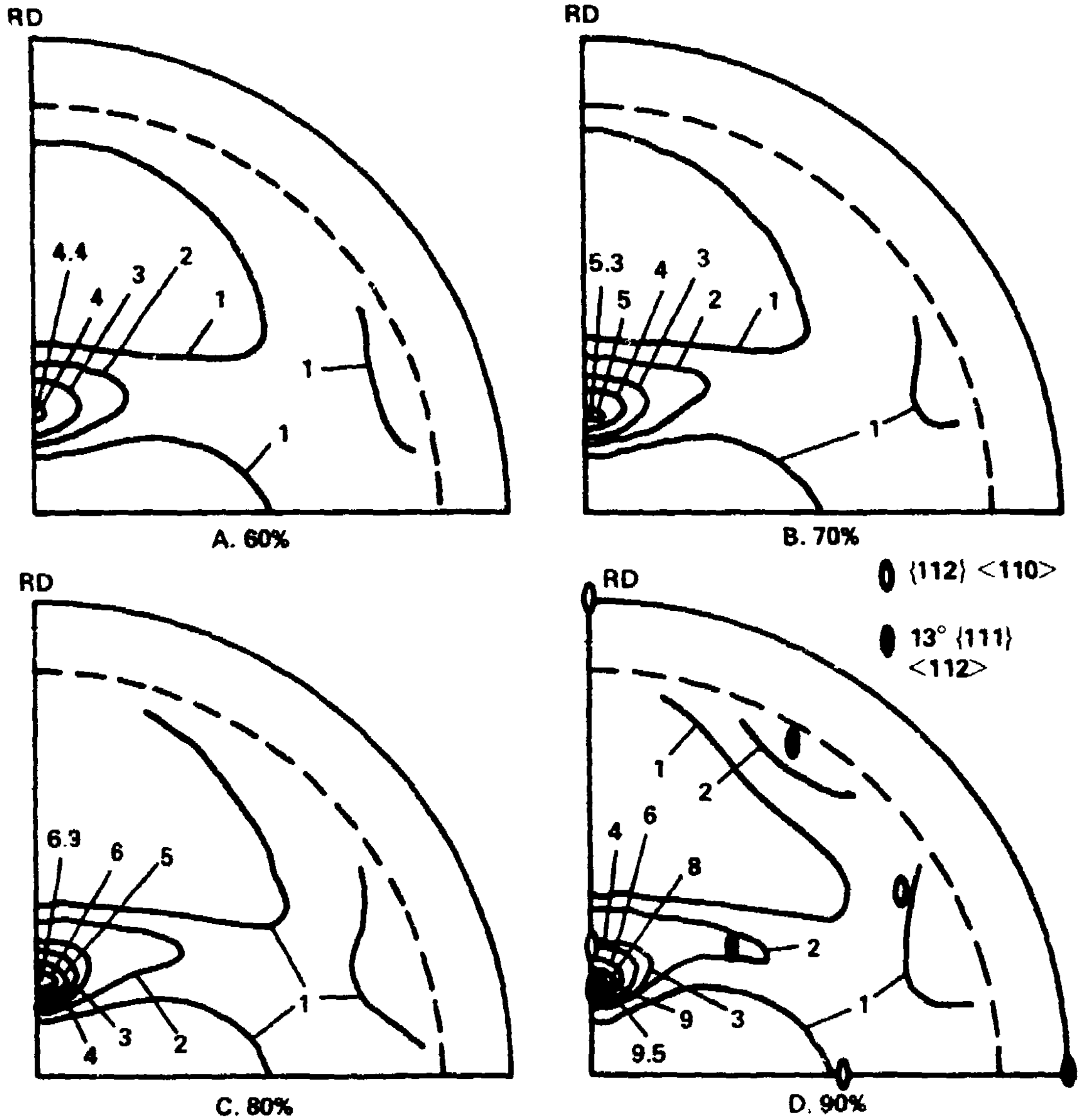
Aging Temperature, °F	Aging Time, hr	Hardness DPH
1400 (760°C)	0	327 (as-rolled)
	1	336
	2	335
	4	313
	8	298
	16	283
	32	283
1300 (704°C)	0	327 (as-rolled)
	1	354
	2	355
	4	360
	8	360
	16	357
	32	356

Table X

Precipitation Hardening of the Specimens Rolled 90 Percent at 1500°F (816°C), Recrystallized at 2000°F (1093°C) for 1 Hour, Then Aged for Various Times at 1300°F (704°C)

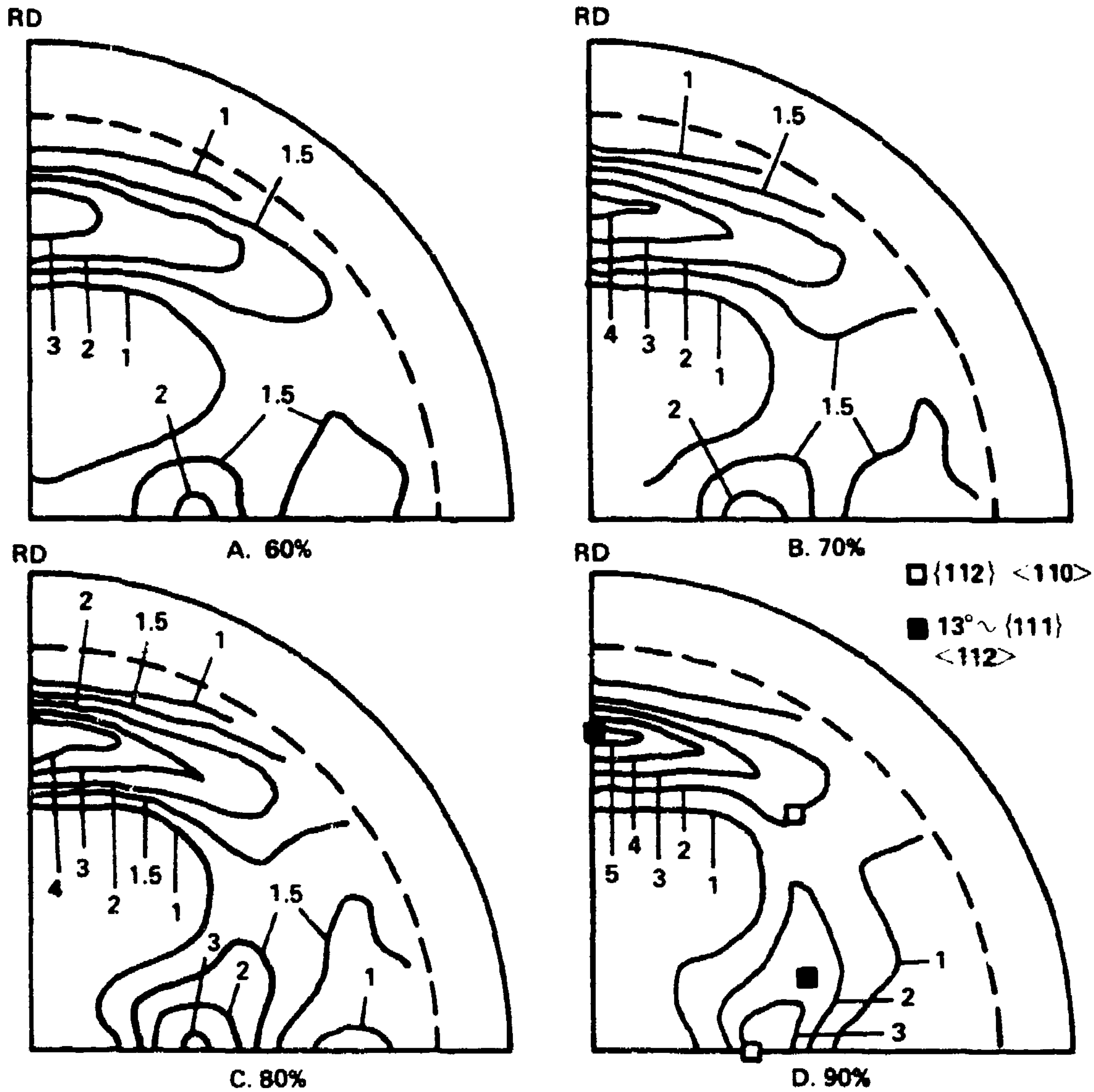
Aging Temperature, °F	Aging Time, hr	Hardness DPH
1300 (704°C)	0	124*
	1	220
	2	242
	4	254
	8	268
	16	280
	32	281

*As-recrystallized



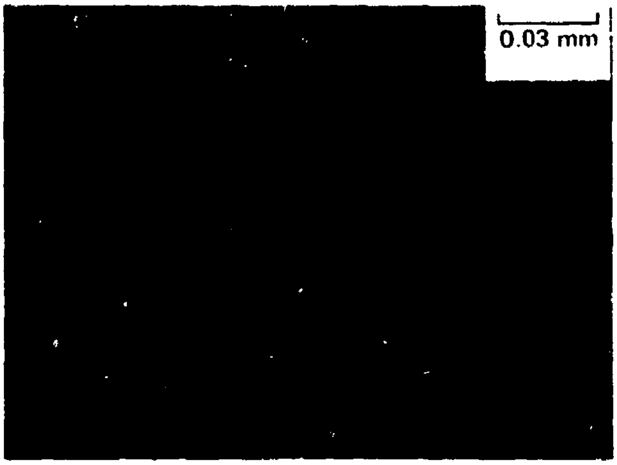
(110) POLE FIGURES OF PLATES ROLLED WITH DECLINING TEMPERATURES TO VARIOUS REDUCTIONS, THEN QUENCHED AND TEMPERED

Figure 1

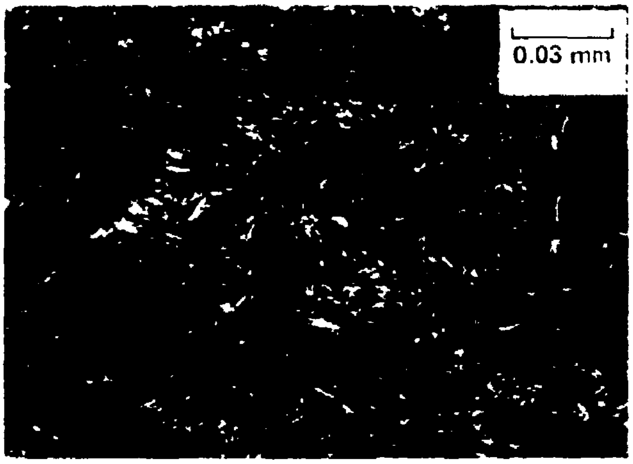


(200) POLE FIGURES OF PLATES ROLLED WITH DECLINING TEMPERATURES TO VARIOUS REDUCTIONS, THEN QUENCHED AND TEMPERED

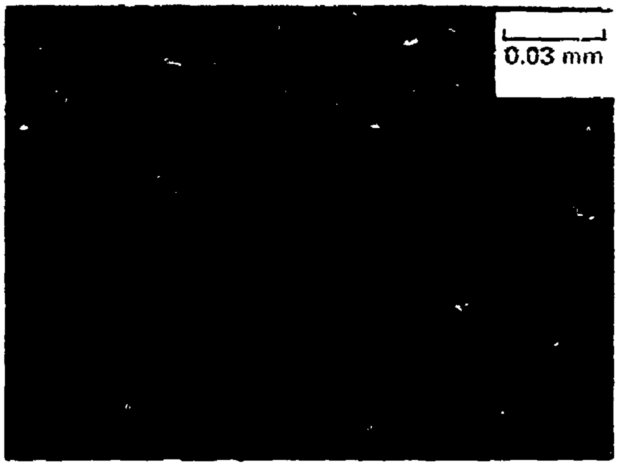
Figure 2



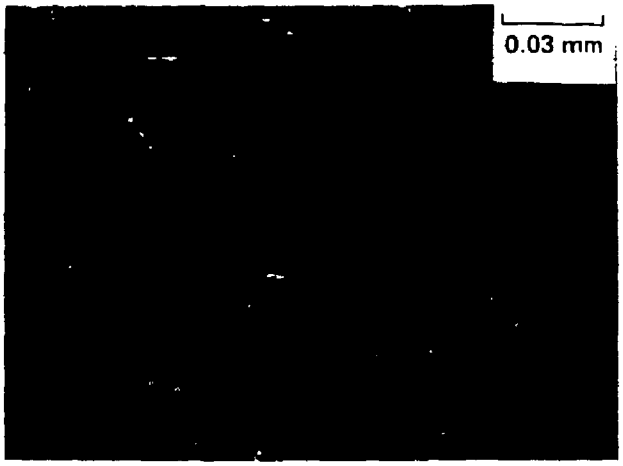
LONGITUDINAL SECTION
ROLLED 60%



TRANSVERSE SECTION
ROLLED 60%



LONGITUDINAL SECTION
ROLLED 90%



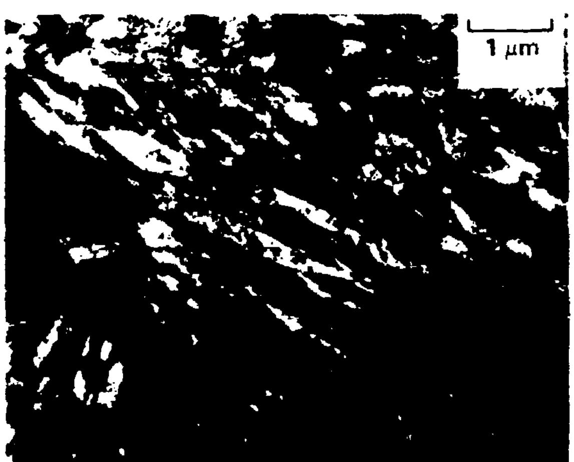
TRANSVERSE SECTION
ROLLED 90%

OPTICAL MICROGRAPHS SHOWING STRUCTURE OF PLATES ROLLED 60 AND 90 PERCENT WITH DECLINING TEMPERATURES, THEN QUENCHED AND TEMPERED. THICKNESS DIRECTION VERTICAL. NITAL ETCH.

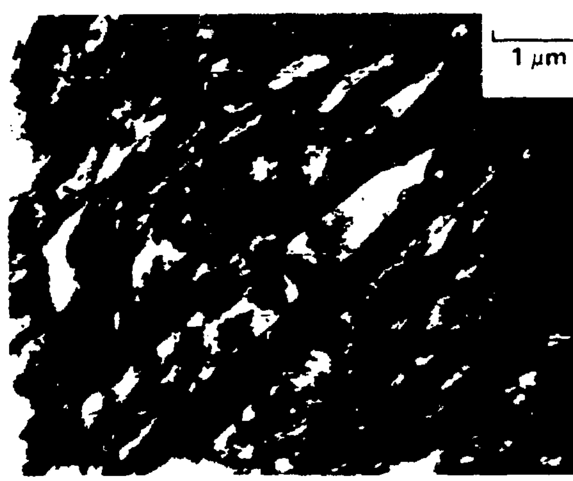
Figure 3

LONGITUDINAL SECTION
ROLLED 60%

TRANSVERSE SECTION
ROLLED 60%



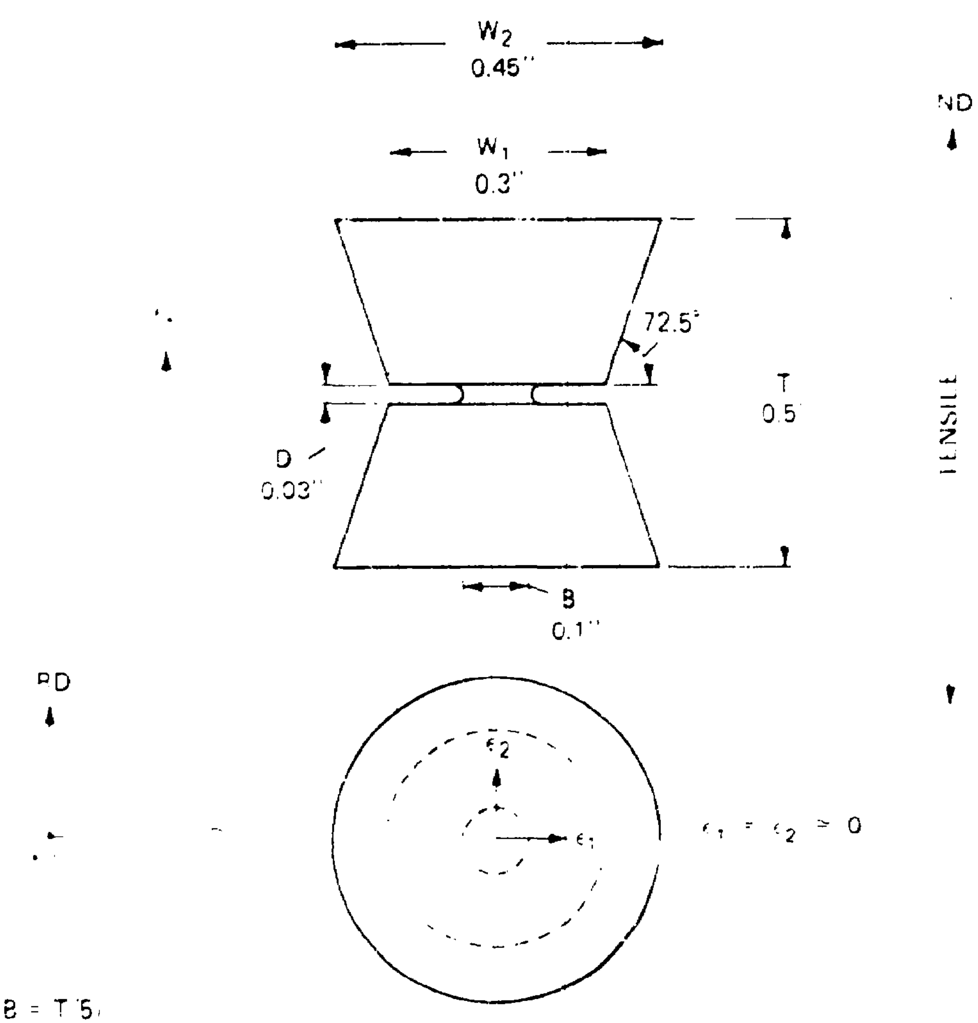
LONGITUDINAL SECTION
ROLLED 90%



TRANSVERSE SECTION
ROLLED 90%

TEM MICROGRAPH SHOWING MARTENSITE STRUCTURE OF PLATES ROLLED 60 AND 90 PERCENT WITH DECLINING TEMPERATURES, THEN QUENCHED AND TEMPERED

Figure 4



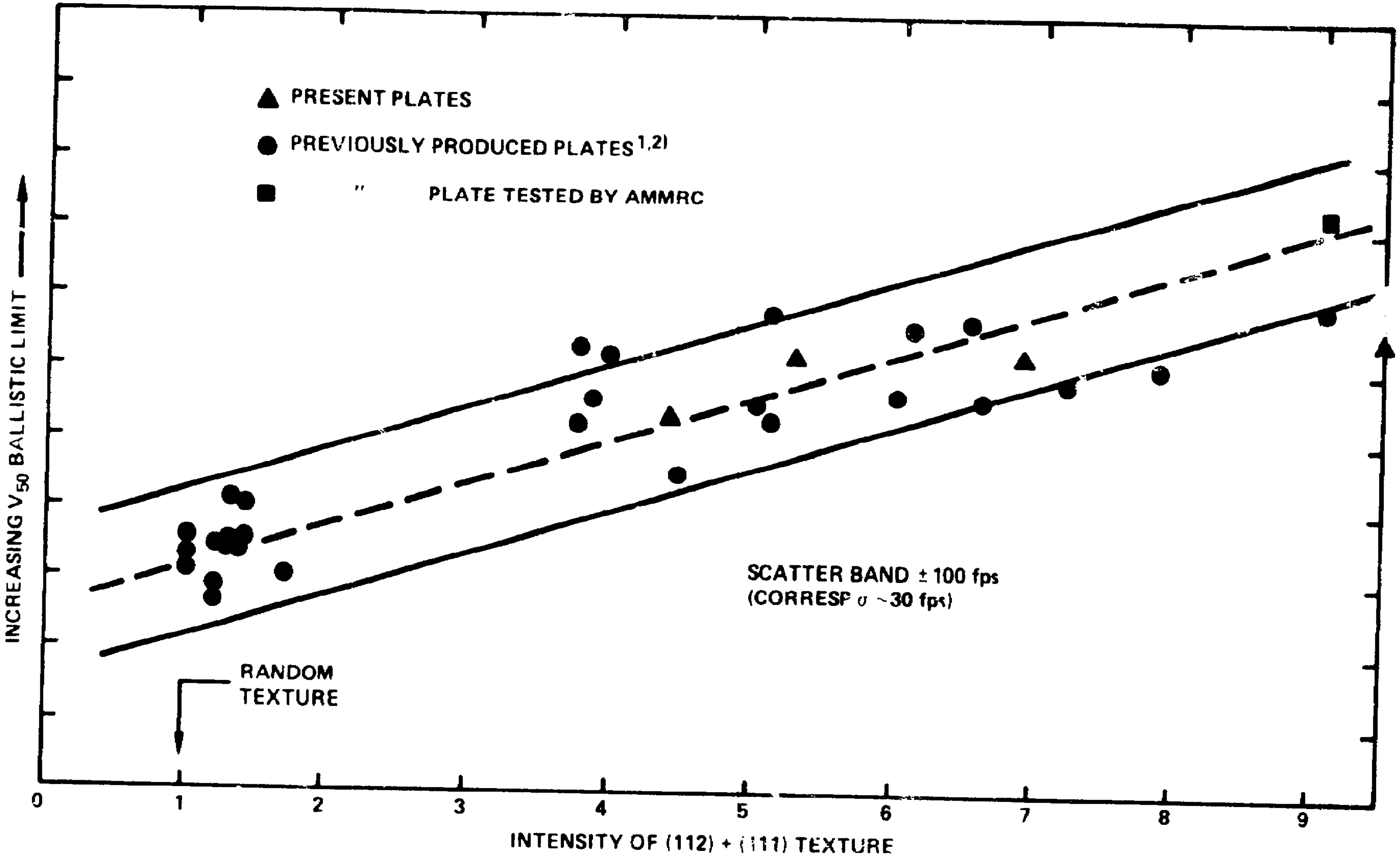
B = W/3, D = B/3.3, B = T/5)

SCALE 4:1

1 inch = 25.4 mm

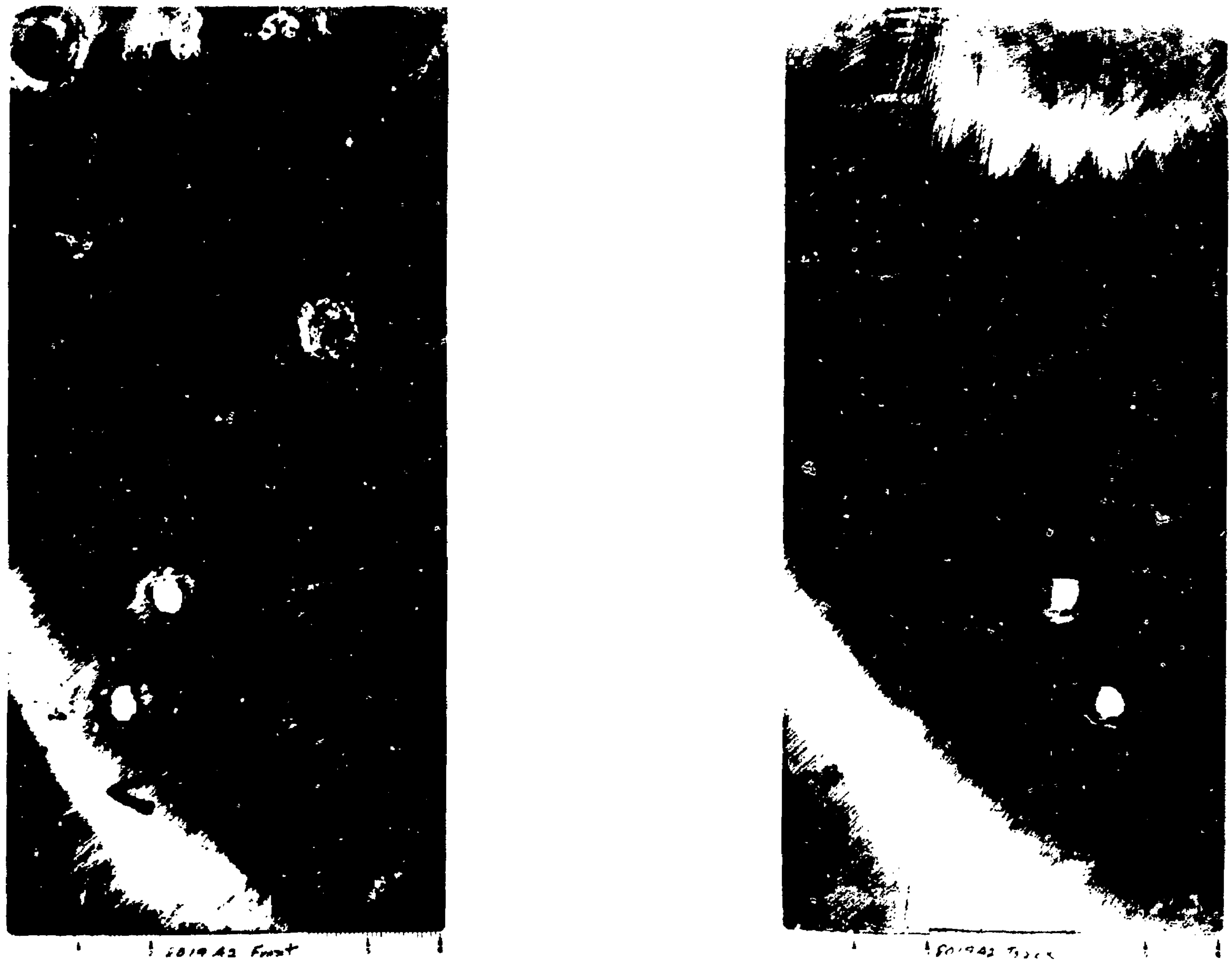
THROUGH-THICKNESS NOTCHED TENSILE SPECIMEN FOR TESTING SPALLING RESISTANCE OF PLATE (STRAIN RATE CONSTANT)

Figure 5



CORRELATION OF THE BALLISTIC LIMIT WITH THE INTENSITY OF (112) + (111) TEXTURE

Figure 6

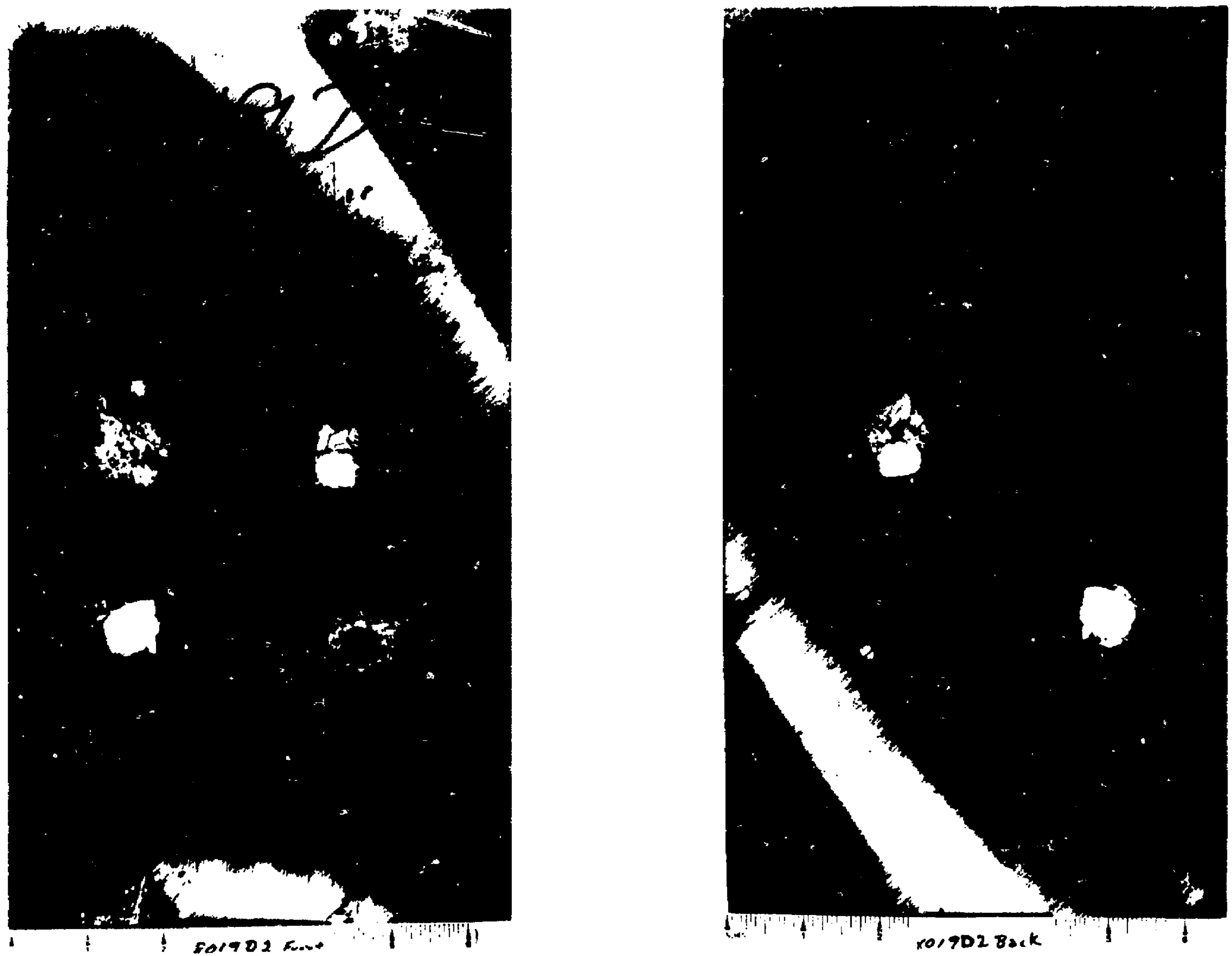


BALLISTIC-TESTED PLATE A-2 ROLLED 60% WITH DECLINING TEMPERATURES THEN QUENCHED AND TEMPERED

Figure 7A

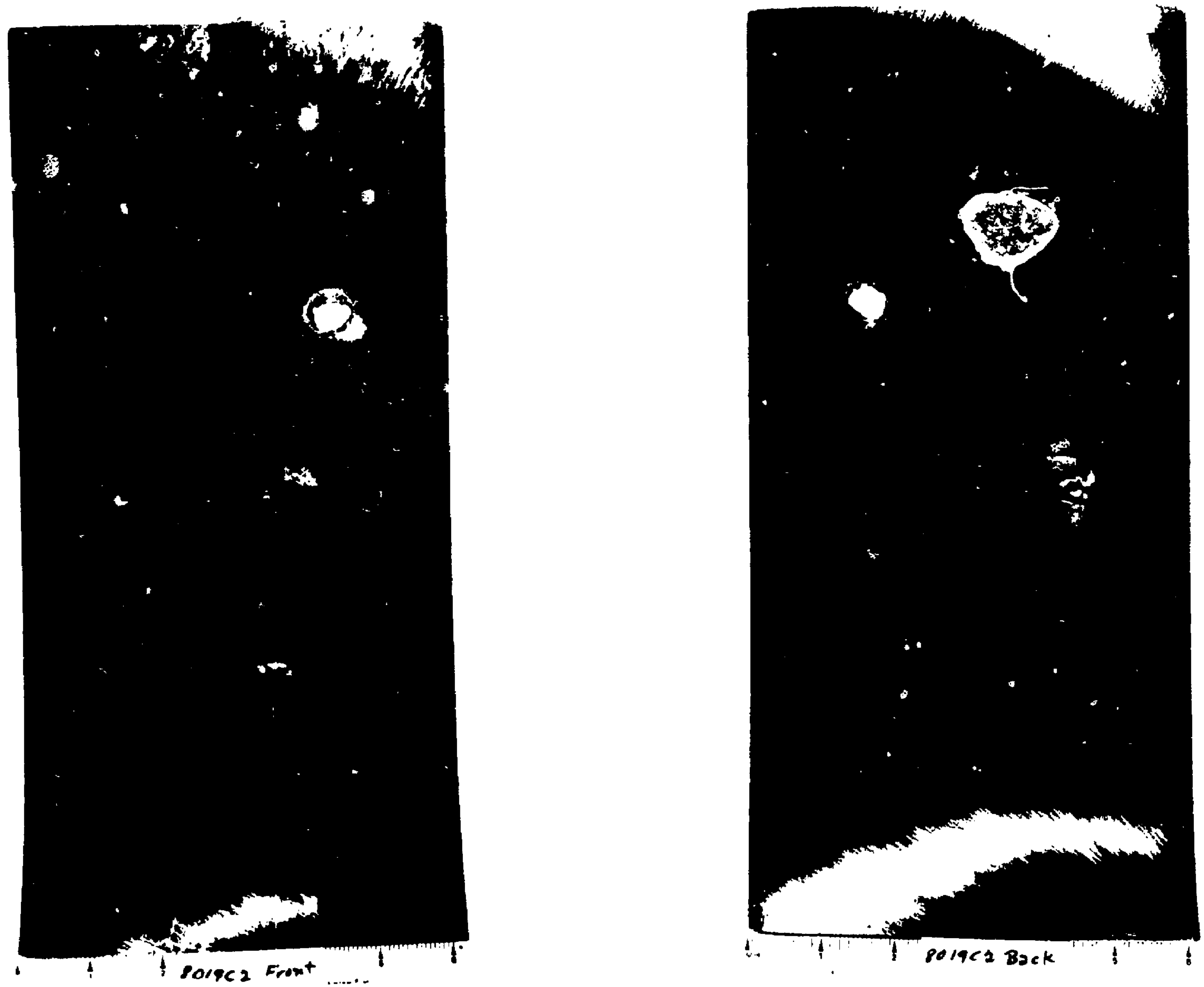
BALLISTIC-TESTED PLATE B-2 ROLLED 70% WITH DECLINING TEMPERATURES THEN QUENCHED AND TEMPERED

Figure 7B



BALLISTIC-TESTED PLATE D-2 ROLLED 90% WITH DECLINING TEMPERATURES THEN QUENCHED AND TEMPERED

Figure 7C



BALLISTIC-TESTED PLATE C-2 ROLLED 80% WITH DECLINING TEMPERATURES THEN QUENCHED AND TEMPERED

Figure 7D

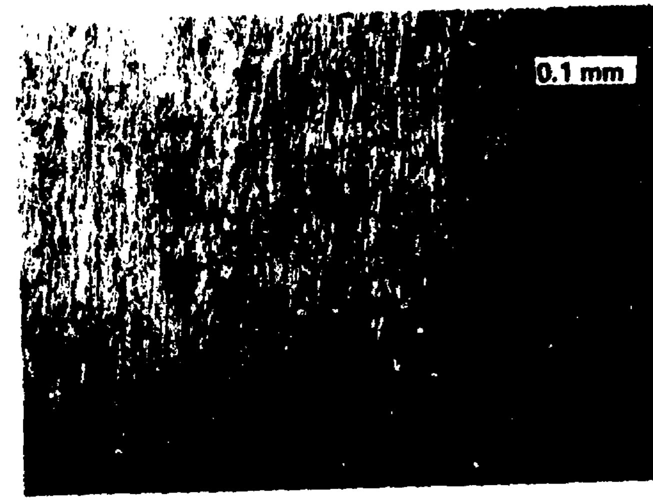
A. 2300°F (1260°C)

B. 1800°F (982°C)

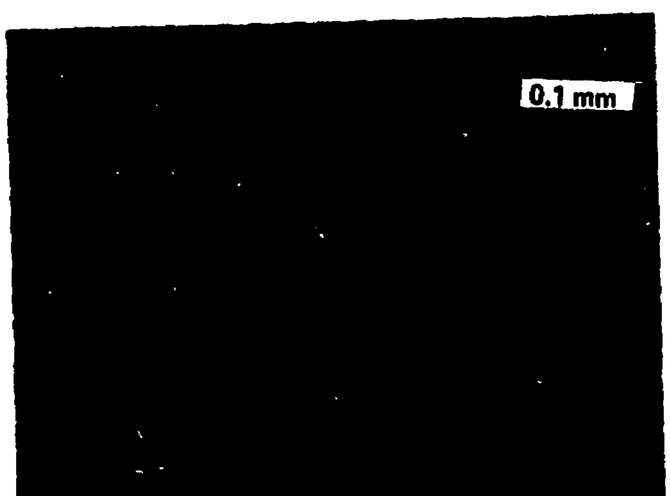
MICROSTRUCTURE OF SPECIMENS ANNEALED ONE HOUR AT TEMPERATURES INDICATED THEN QUENCHED. TRANSVERSE SECTION. HYDROCHLORIC-NITRIC-ACETIC ETCH.

Figure 8

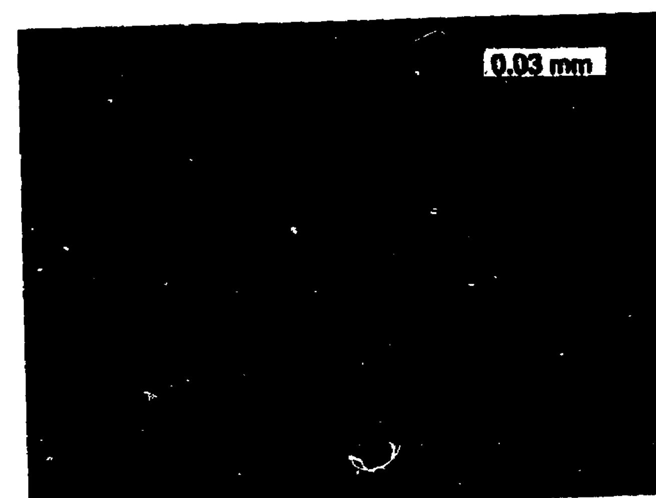
A. 1700°F (927°C)



B. 1500°F (816°C)



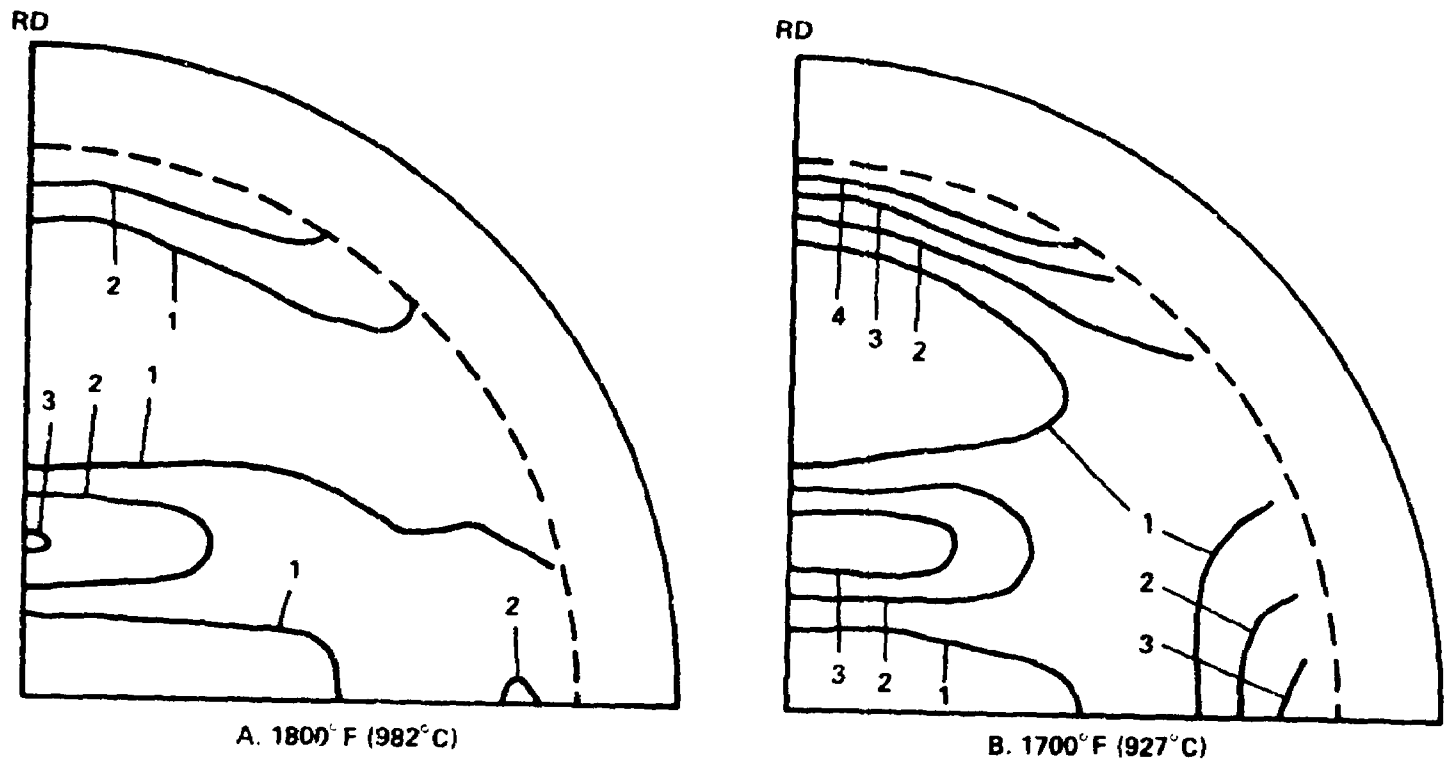
C. 900°F (482°C)



D. -100°F (-73°C)

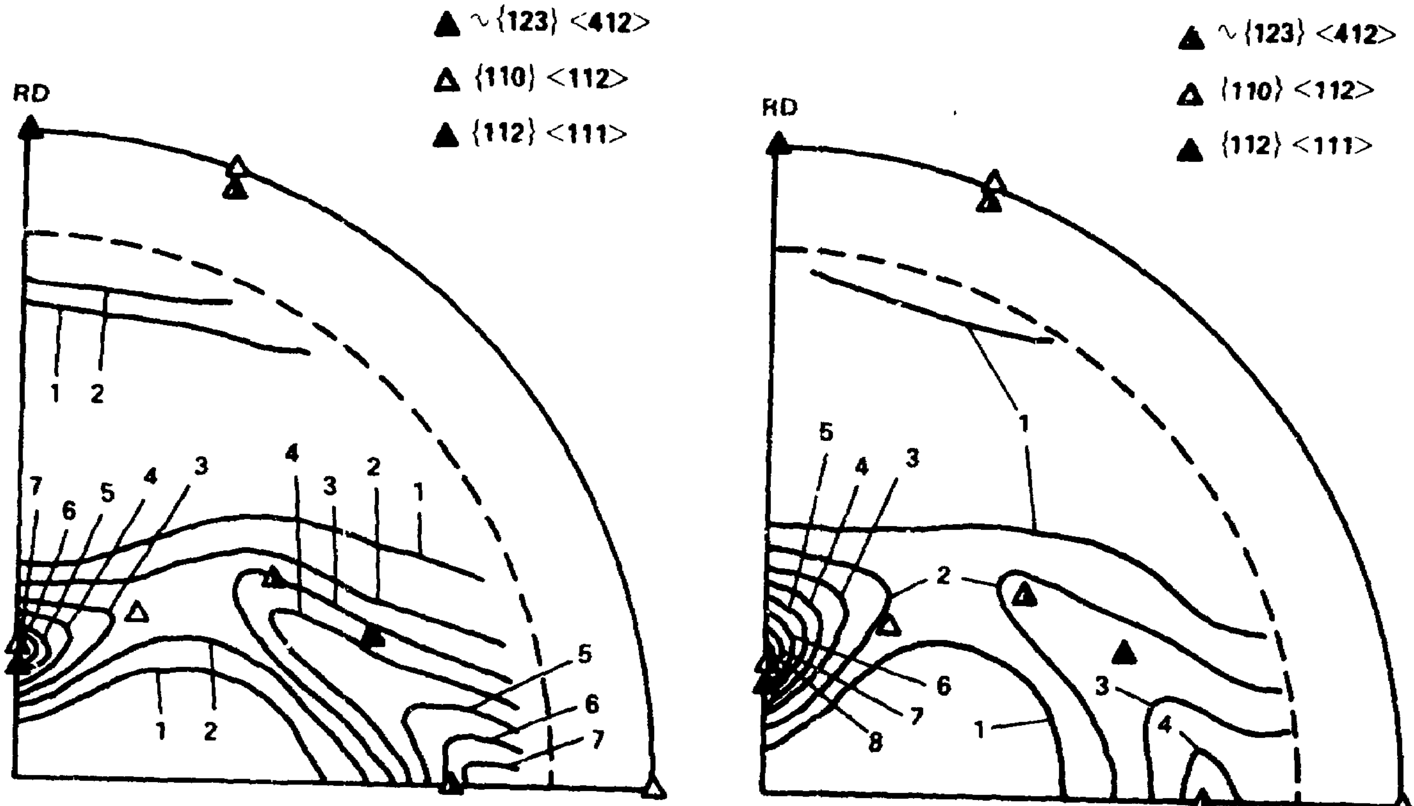
MICROSTRUCTURE OF SPECIMENS SOLUTION TREATED AT 1800°F (982°C) THEN ROLLED 90% AT TEMPERATURES INDICATED. TRANSVERSE SECTION. HYDROCHLORIC-NITRIC-ACETIC ETCH.

Figure 9



(111) POLE FIGURES OF SPECIMENS ISOTHERMALLY ROLLED 90% AT 1800°F (982°C) AND 1700°F (927°C)

Figure 10

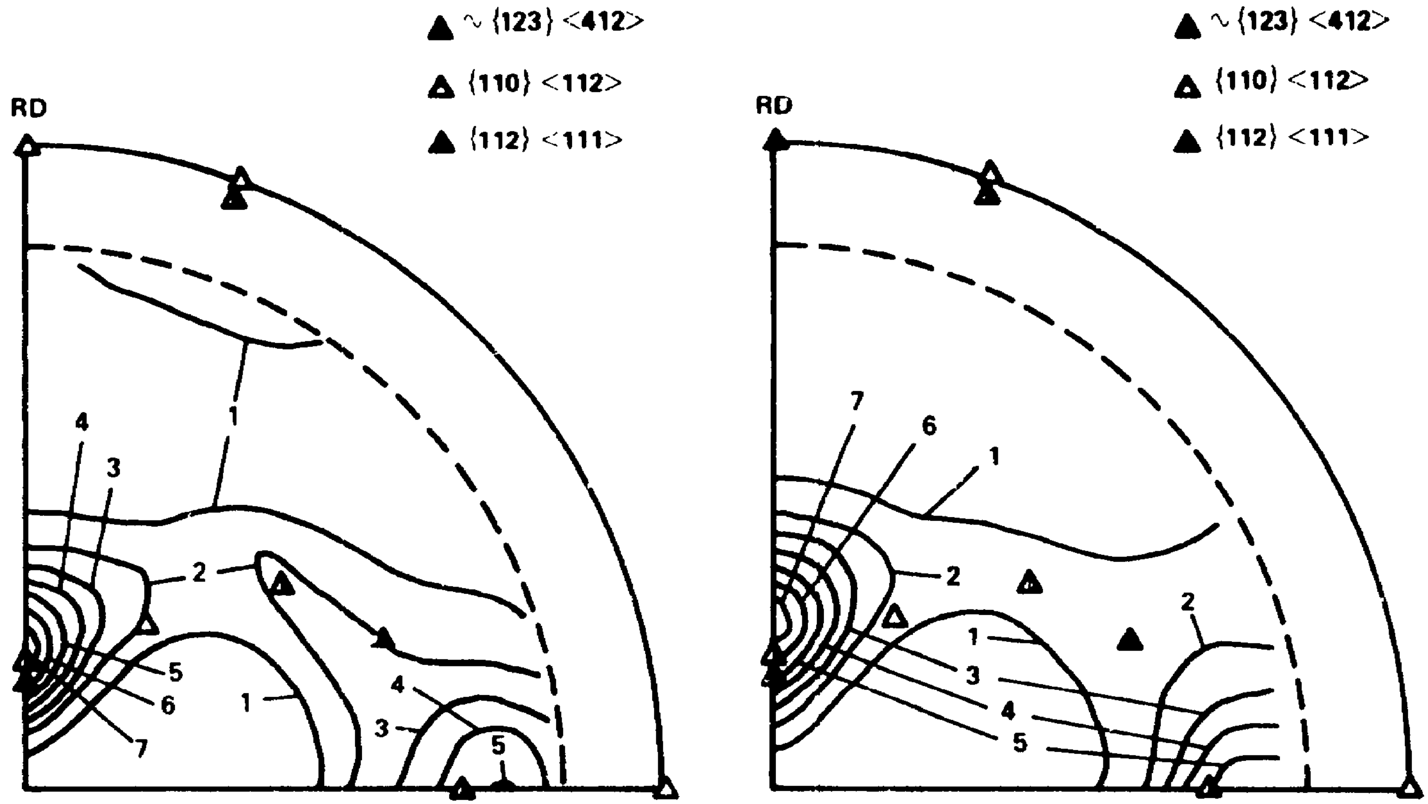


A. 1500°F (816°C)

B. 900°F (482°C)

{111} POLE FIGURES OF SPECIMENS ISOTHERMALLY ROLLED 90% AT 1500°F (816°C) AND 900°F (482°C)

Figure 11

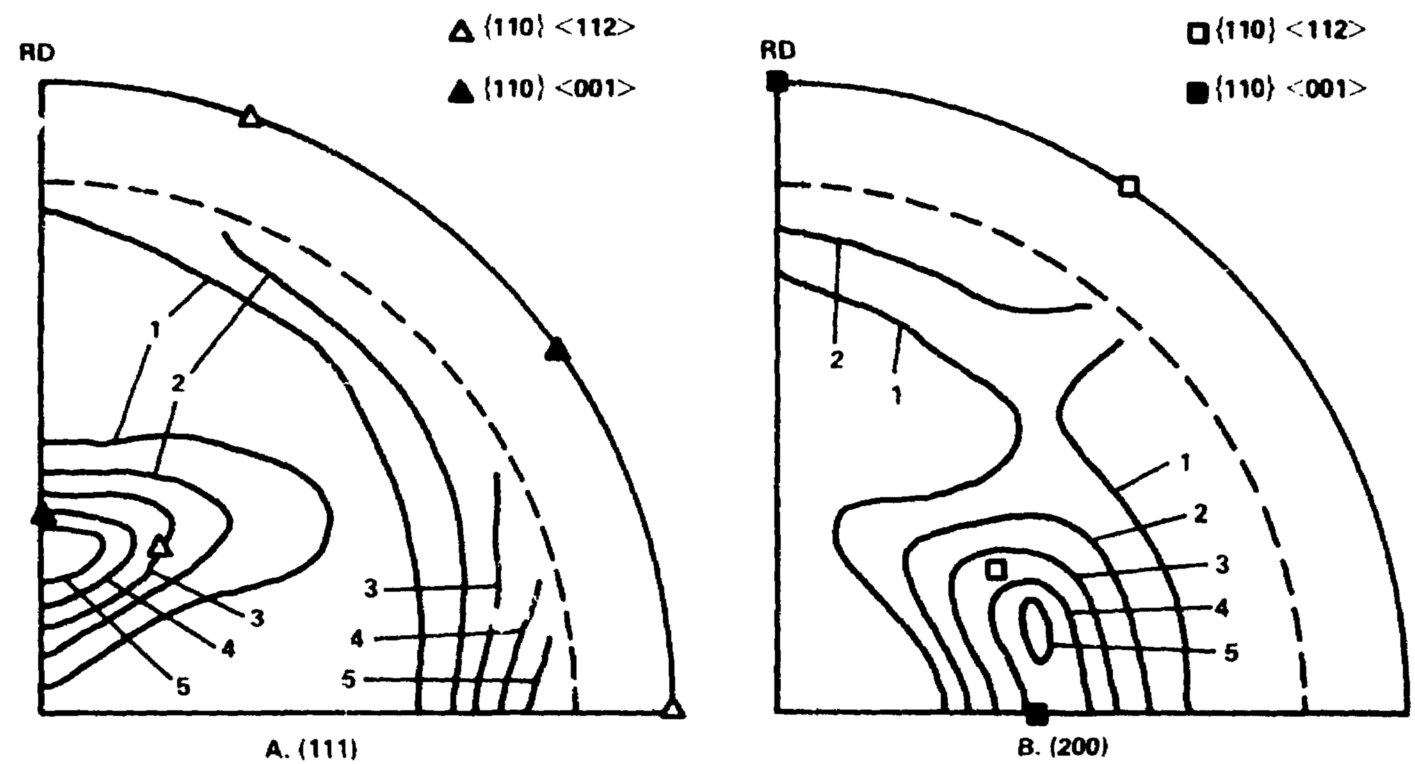


A. 700°F (371°C)

B. 500°F (260°C)

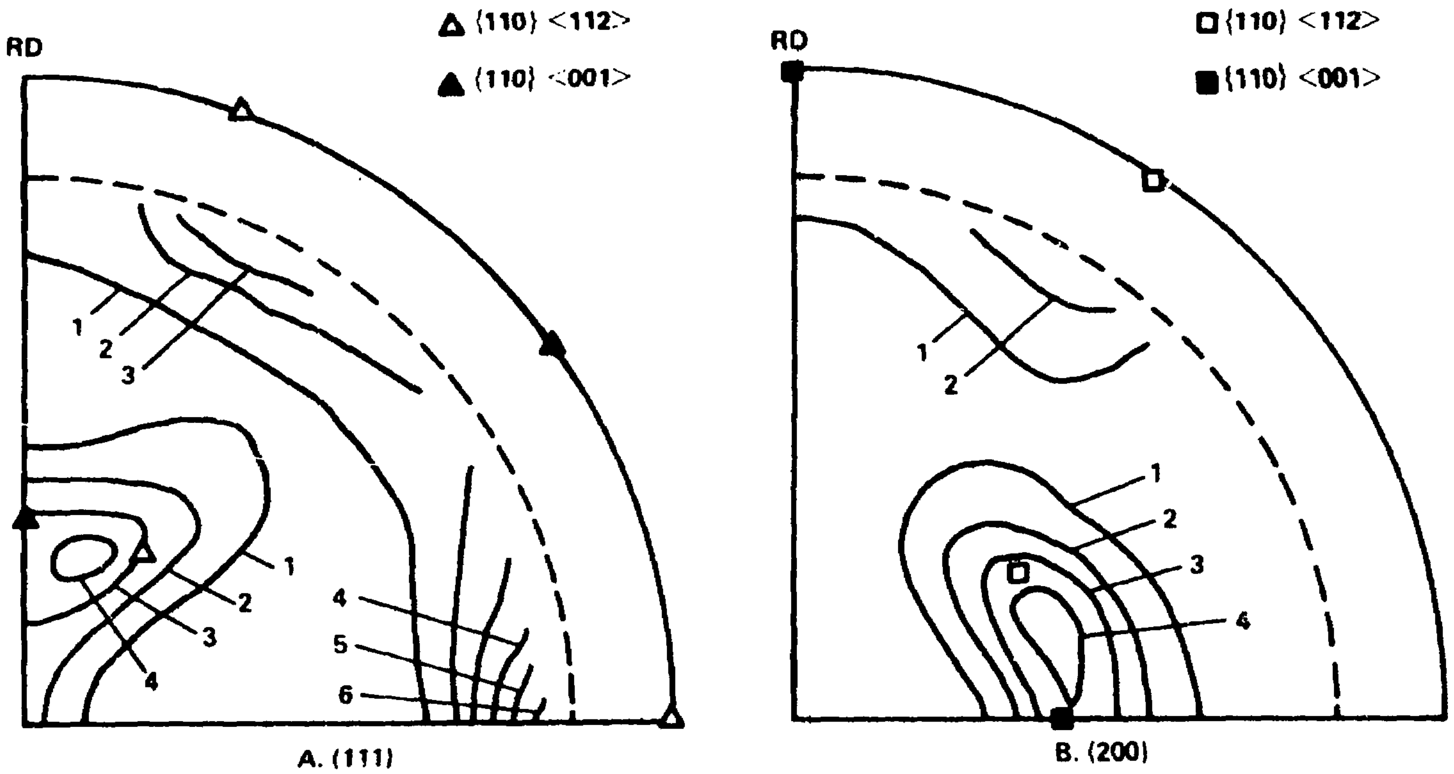
(111) POLE FIGURES OF SPECIMENS ISOTHERMALLY ROLLED 90% AT 700°F (371°C) AND 500°F (260°C)

Figure 12



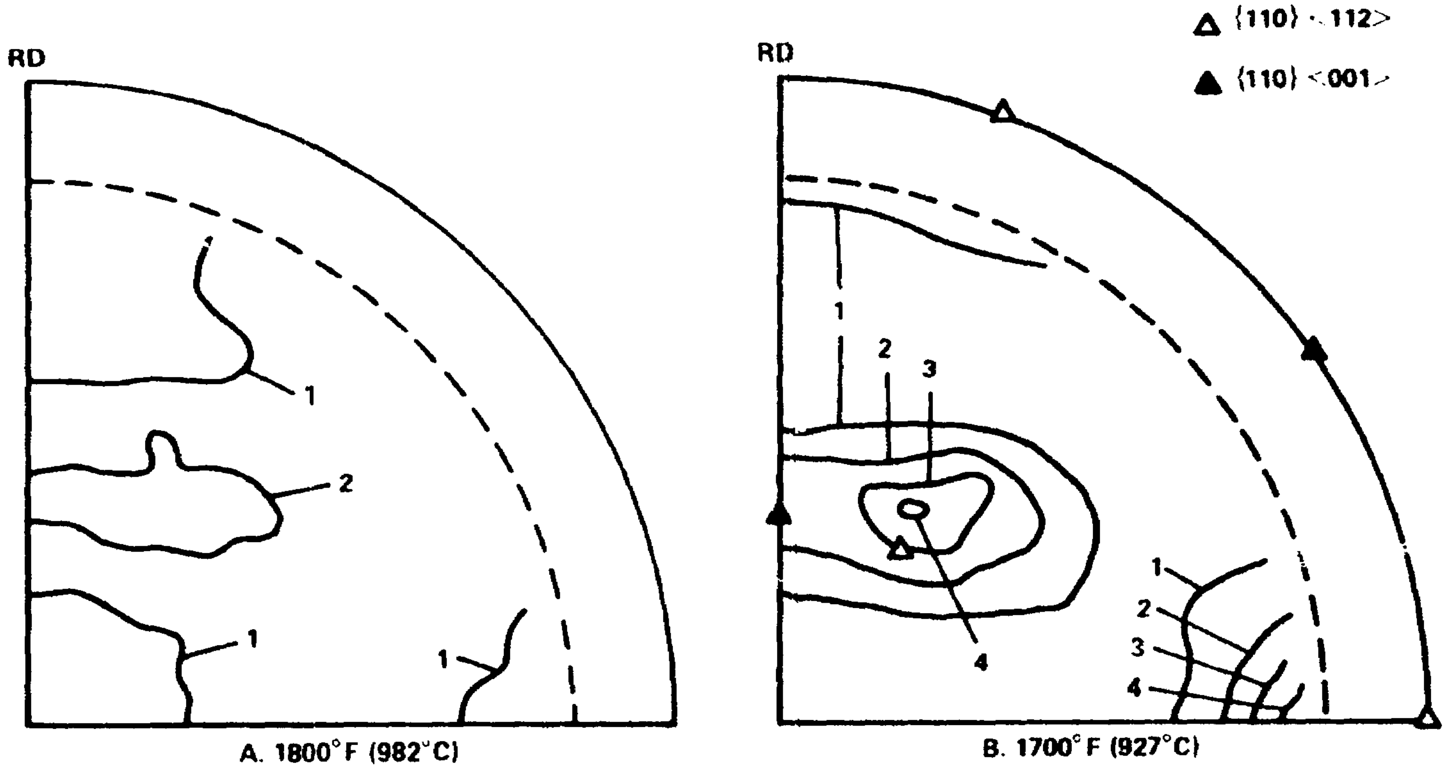
(111) AND (200) POLE FIGURES OF SPECIMEN ISOTHERMALLY ROLLED 90% AT 80°F (27°C)

Figure 13



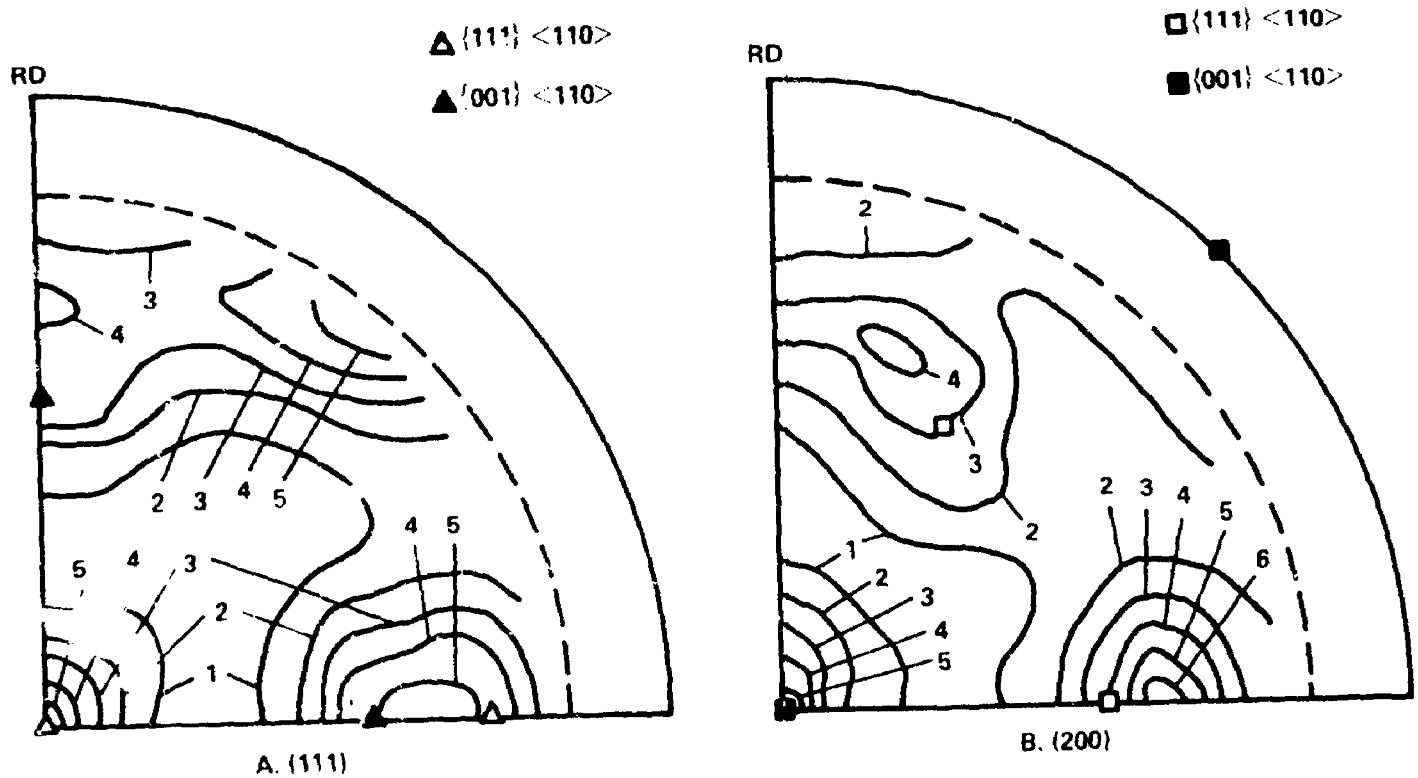
(111) AND (200) POLE FIGURES OF SPECIMEN ISOTHERMALLY ROLLED 90% AT -100°F (-73°C)

Figure 14



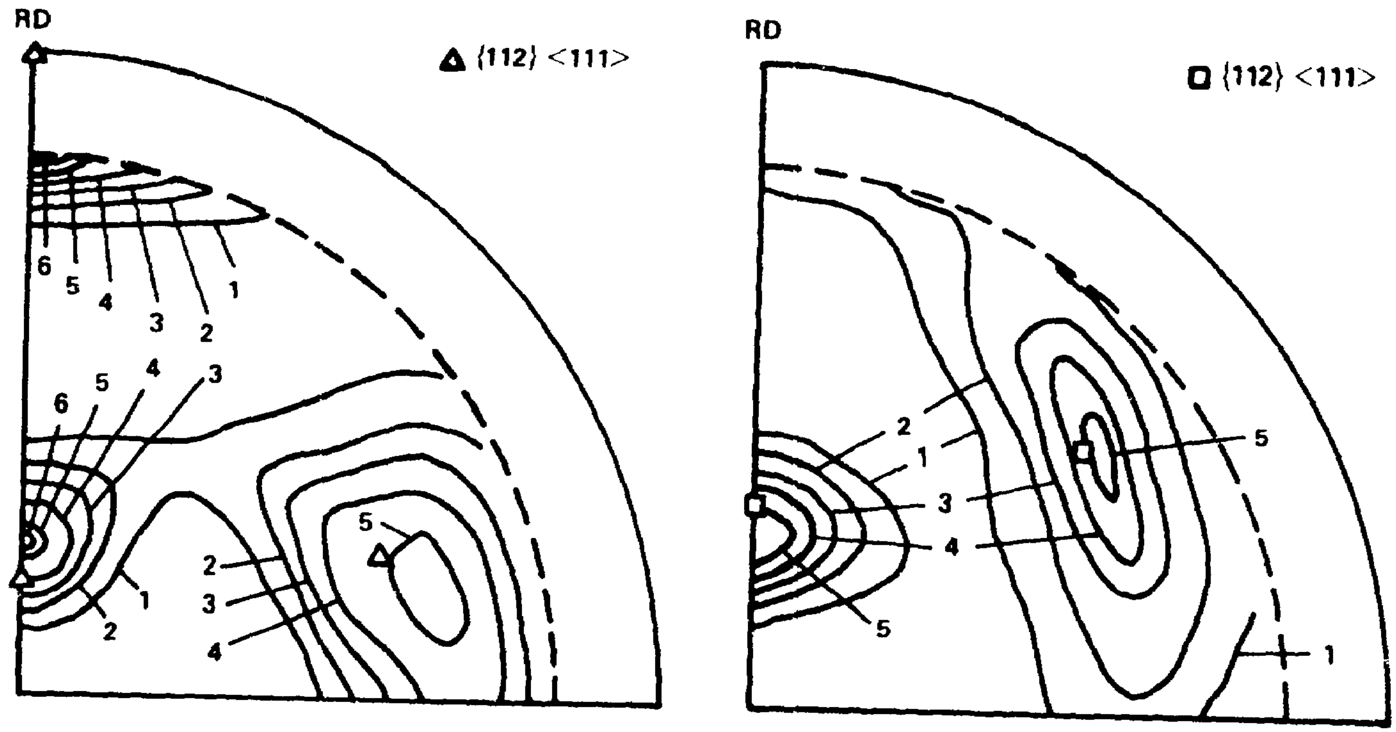
(111) POLE FIGURES OF SPECIMENS ISOTHERMALLY ROLLED 90% AT 1800°F (982°C) AND 1700°F (927°C), THEN ANNEALED AT 2000°F (1093°C) FOR ONE HOUR

Figure 15



(111) AND (200) POLE FIGURES OF SPECIMEN ISOTHERMALLY ROLLED 90% AT 1500°F (816°C), THEN ANNEALED AT 2000°F (1093°C) FOR ONE HOUR

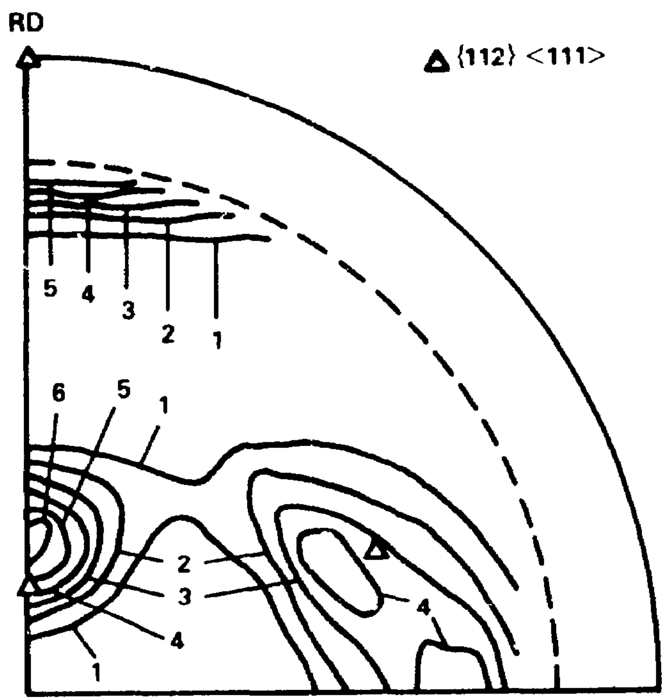
Figure 16



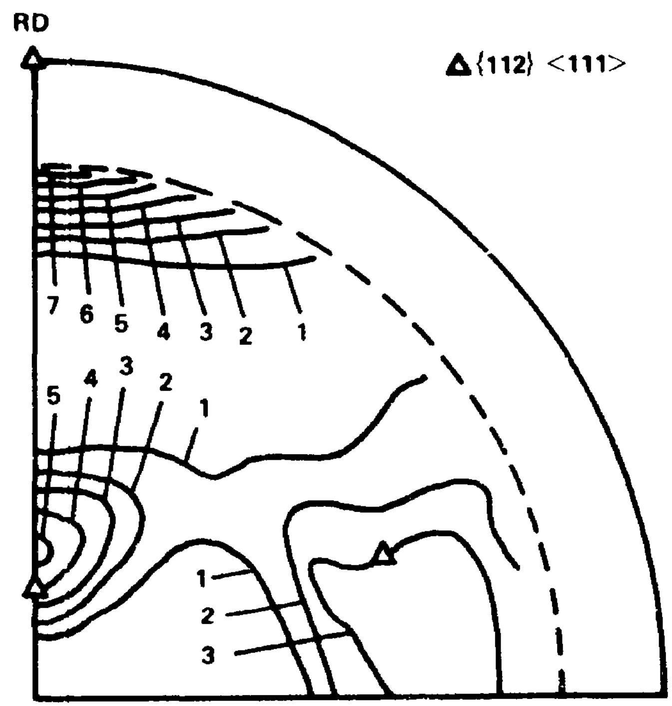
A. (111) B. (200)

(111) AND (200) POLE FIGURES OF SPECIMEN ISOTHERMALLY ROLLED 90% AT 900°F (482°C), THEN ANNEALED AT 1800°F (982°C) FOR ONE HOUR

Figure 17



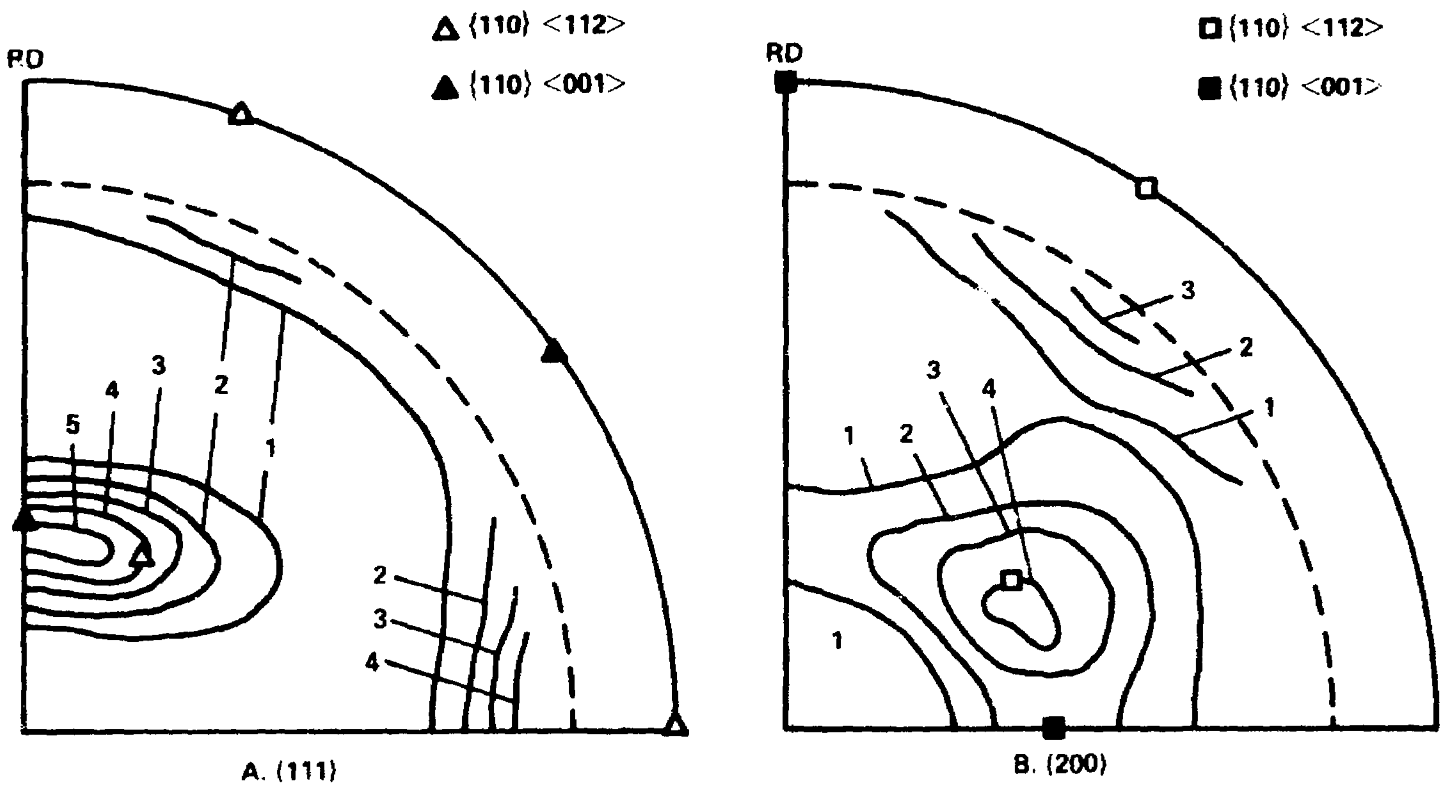
A. 700°F (371°C)



B. 500°F (260°C)

(111) POLE FIGURES OF SPECIMENS ISOTHERMALLY ROLLED 90% AT 700°F (371°C) AND 500°F (260°C), THEN ANNEALED AT 2000°F (1093°C) FOR ONE HOUR

Figure 18



(111) AND (200) POLE FIGURES OF SPECIMEN ISOTHERMALLY ROLLED 90% AT -100°F (-73°C), THEN ANNEALED AT 1800°F (982°C) FOR ONE HOUR

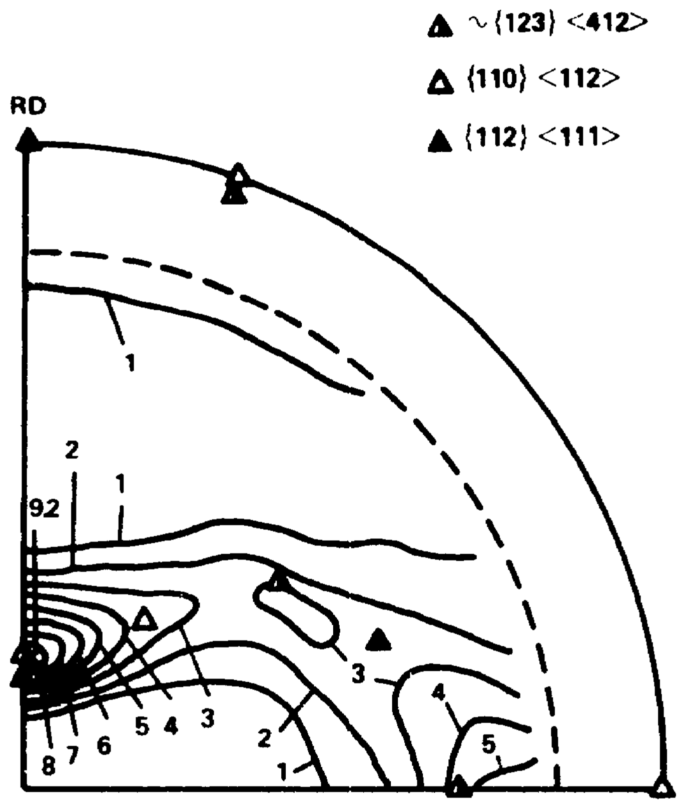
Figure 19

A

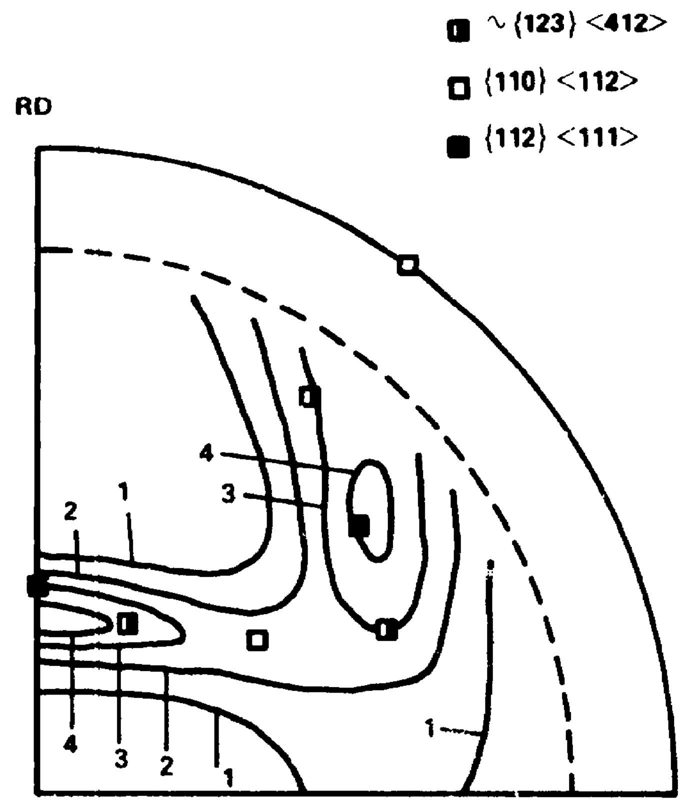
B

MICROSTRUCTURES OF SPECIMENS: A. ROLLED 90% AT 1500°F (816°C) AND RECRYSTALLIZED AT 2000°F (1093°C); B. SAME AS IN A AND AGED AT 1300°F (704°C) FOR 16 HOURS. TRANSVERSE SECTION. HYDROCHLORIC-NITRIC-ACETIC ETCH.

Figure 20



A. (111)



B. (200)

(111) AND (200) POLE FIGURES OF THE B-1 PLATE OF THE A-286 ALLOY, ROLLED 70% AT TEMPERATURES FROM 1550 TO 1380°F (843 TO 749°C)

Figure 21

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The scope of the present research included two parts: (1) to further strengthen the (112)+(111) texture in the 5Ni steel (actually a medium-carbon 5Ni-Si-Cu-Mo-V steel) armor by introducing modifications to the previously employed hot-rolling process, and to provide latitude in production of this superior textured armor plate; (2) to investigate the crystallographic texture formation in an austenitic steel system (such as the

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ABSTRACT

A-286 alloy) so that textures other than those formed in quenched-and-tempered martensite can be produced and their effects on properties can be studied.

Results indicate that the intensity of the (112)+(111) texture in the 5Ni steel armor can be further increased by controlled rolling with declining temperatures, and that the V_{50} ballistic limits of the 60, 70, and 80 percent rolled plates correlate with the texture-intensity parameter within the scatter band of the earlier data. However, for the 90 percent rolled plates, the V_{50} ballistic limit falls below the scatter band of the correlation. Also, the back-spalling resistance of the heavily rolled plates appears to be lower. These results have been discussed in relation to the changes of other mechanical properties as a consequence of the low rolling and finishing temperatures employed.

Results from the present study on the A-286 alloy indicate that a complete transition of the rolling texture from the copper-type to the brass-type can be effected by decreasing the rolling temperature. A large variety of textures can be produced by appropriate thermal and mechanical processing treatments. These textures can probably be utilized to advantage only in specific applications for sheet materials. The alloy does not appear to be suitable for producing strongly textured thick plates because of the very limited temperature range for hot working to high reductions without concurrent recrystallization.

Army Materials and Mechanics Research Center,
Watertown, Massachusetts 02172
EFFECT OF CONTROLLED ROLLING ON THE CRYSTALLOGRAPHIC TEXTURE AND THE MECHANICAL AND BALLISTIC PROPERTIES OF STEEL ARMOR PLATES
Koun Wu
Final Technical Report AMMRC TR 80-6, March 1980, 47 pp - illus.
DAAG Project 46-79-C-0010, AMMRC Code [illegible], Unclassified Report

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